Transactions

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TRANSACTIONS

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Heat Treatment — Its Effect On the Combined Carbon and Physical Properties of Wheel Iron

By S. C. Massari*, Chicago, Ill.

Abstract

The heat treatment of cast iron is growing in importance and its successful accomplishment is largely the result of a thorough understanding of the behavior of the metal when subjected to various heat treating temperatures. Whether this heat treatment involves merely annealing or reheating to various temperatures prior to quenching, the final result is influenced by the temperature and the time at the temperature to which the casting is subjected. This investigation describes the effect of heat treating a low silicon gray cast iron typical of the plate and hub of a chilled car wheel. The work involves the heat treatment of a long series of specimens at temperatures ranging from 500° F. to a maximum of 1700° F. for various lengths of time to a maximum of 96 hr. and the subsequent determination of the important physical properties of the metal.

Introduction

1. Chilled car wheels, after being shaken from the mold, are annealed by controlled cooling in annealing pits, utilizing the sensible heat within the wheel. Such a heat treatment is designed to relieve all internal casting stresses, and in addition, accomplish certain structural changes, especially in the chilled portions. It is important that this heat treatment be so conducted that there is minimum loss in physical properties in the plate of the wheel. It is largely for this reason that the following described investigation was conducted.

2. Chilled car wheel iron has the approximate composition shown in Table 1.

* Metallurgist, Research Laboratory, Association of Manufacturers of Chilled Car Wheels.

Note: Presented at a session on Gray Iron at the 43rd Annual A.F.A. Convention, Cincinnati, Ohio, May 18, 1939.

3. Metal of this composition when cast into a car wheel mold produces white iron in the tread and flange of the wheel because of the chiller, but the remainder of the casting, which includes the plate and hub of the wheel, is gray iron. Due to the thickness of the section involved and the chemical balance of the metal, the plate and hub of the wheel in the as-cast condition are invariably a gray iron having a combined carbon content of about 0.85 per cent, with flake graphite of normal distribution and size in a matrix composed of lamellar pearlite and practically devoid of free carbides or ferrite. It is the behavior of this particular metal with which this investigation is concerned.

Table 1

APPROXIMATE COMPOSITION OF CHILLED CAR WHEEL IRON

	Per Cent
Silicon	0.50 to 0.60
Manganese	0.50 to 0.60
Total Carbon	3.45 to 3.65
Sulphur	0.14 max.
Phosphorus	0.35 max.

4. It is very important to understand that this research covers only a low silicon iron and the data obtained cannot be directly translated to a high silicon iron, such as automotive cylinder blocks and the like. The high silicon content of such irons exerts a pronounced influence on the behavior of the metal during heat treatment, increasing the rate of combined carbon decomposition during heat treatment and markedly modifying the iron-carbon-silicon equilibrium.

5. The physical properties of iron of a given composition are the result of two principal factors, namely, the graphite and the structure of the matrix. The total quantity of graphite, its character, size and distribution are probably the most important, but with a given graphite structure the structural nature of the matrix is the only other important consideration. During this investigation the iron was cast under identical conditions in order to produce as nearly as possible a constant graphite structure so that the effect of heat treatment on the physical properties could be definitely correlated with the structural changes in the matrix of the iron.

TEST PROCEDURE

6. It should be evident that it was impractical to cut the large number of test specimens involved in the investigation from the plates of car wheels. The labor and time consumed would be far in excess of any reasonable value. It, however, was important that the test results should conform as closely as possible to the behavior to be expected of the metal in the plate and hub of a wheel. Consequently, test slabs of ordinary cupola melted wheel iron were cast of dimensions comparable with the thickness of the metal in the plate of a wheel. These test slabs were 10x12x1½ in. cast in a green sand mold. Micro-examination and chemical analysis definitely revealed that the structure and the combined carbon content in these slabs was identical with that obtained in the as-cast condition in the plate of a chilled car wheel.

7. The cooling rate of these slabs closely approximated the rate at which the metal in the plate of a wheel solidifies. Control analyses run on each group of slabs consistently indicated combined carbon contents ranging from 0.85 to 0.90 per cent. Usually seven of these slabs were cast from the same ladle or consecutive ladles so that the iron would be as nearly identical as possible. The slabs were permitted to remain in the mold over night. They were then removed and cut into pieces of such size that conventional test specimens could be machined from them after heat treatment.

8. One slab of each group was used for control purposes. A complete set of test specimens was machined from the slab without receiving any prior heat treatment. In other words, the physical properties so obtained could be considered as representative of the properties the metal possessed in the as-cast condition. The remaining slabs were placed in an electric furnace, which was automatically controlled, and held at temperature for various lengths of time, ranging from 4 to 96 hr.

9. The heat treatment temperatures used in the investigation and the chemical analysis of the control slab of the group treated at each temperature are shown in Table 2.

10. A group was held at each temperature for 8, 24, 48, 72 and 96 hr. At the end of each of these periods a slab was removed from the furnace and buried in Sil-O-Cel to restrain the rate of cooling to room temperature.

11. Tests were made on specimens cut from each slab to determine the tensile, compressive, transverse, impact and hardness properties. In addition, chemical analyses were made to determine the amount of residual combined carbon. The test results throughout the investigation were in close agreement, and the values reported can be considered extremely reliable because of this close reproducibility. In most instances each value reported represents an average of two tests and at times, when the test results appeared to be erroneous, because of a lack of thorough understanding of the behavior of the metal, the tests were repeated four times.

Table 2

HEAT TREATMENT TEMPERATURES AND CHEMICAL ANALYSES OF CONTROL SLABS

Heat Treatment Temp.	*Chemical Analysis*						
°F.	*T.C.*	*G.C.*	*C.C.*	*Si*	*Mn*	*P*	*S*
500	3.46	2.68	0.78	0.52	0.55	0.308	0.128
600	3.47	2.69	0.78	0.44	0.52	0.294	0.134
700	3.44	2.59	0.85	0.48	0.50	0.310	0.114
800	3.53	2.75	0.78	0.48	0.50	0.292	0.138
900	3.54	2.79	0.75	0.48	0.52	0.300	0.128
1000	3.49	2.69	0.80	0.48	0.50	0.280	0.125
1100	3.54	2.76	0.78	0.46	0.49	0.285	0.125
1200	3.57	2.73	0.84	0.51	0.57	0.286	0.117
1310	3.51	2.70	0.81	0.51	0.58	0.310	0.132
1400	3.54	2.78	0.76	0.52	0.57	0.297	0.122
1500	3.62	2.78	0.84	0.50	0.52	0.307	0.107
1600	3.50	2.60	0.90	0.53	0.53	0.300	0.119
1700	3.41	2.60	0.81	0.49	0.52	0.308	0.135

12. Tension tests were made according to A.S.T.M. standard specification A48-36 for gray iron. The bar used is known as test bar *"B."* This bar has a nominal diameter of 0.800 in. with a straight section 1 in. long. Tensile tests in which elongations were measured during the time the load was being applied were made on a bar of identical design except that the straight portion of the bar was increased to 2 in. so that an extensometer could be applied to a 2 in. gauge length. The tension tests were made in a Riehle 100,000 lb. Universal testing machine equipped with spherical heads to permit self-alignment of the bar and thereby eliminate bending stresses as much as possible.

13. Compression test specimens were cylinders having a diameter of 1.1285 in. (1 sq. in. cross sectional area) and a length of 3 in. The compression specimens were likewise tested in the same testing machine using an extensometer on a 2 in. gauge length.

14. The transverse tests were made on a bar 12 in. long, 1 in. sq. and tested on supports 10 in. apart. Deflection of the bar was measured by means of an Ames dial reading in thousandths of inches. The dial was so held that the spindle was resting against the knife edge at the center of the bar.

15. Impact tests were run in a Riehle impact testing machine with the pendulum delivering a blow of 110 ft. lb. The impact bar was 1 in. sq., unnotched, and freely supported on a 5 in. span.

OBSERVED DATA

16. When the data was collected it became apparent that the behavior of the iron could be classified according to three temperature ranges, namely:

(A) Heat treatment below 1000°F.
(B) Heat treatment between 1000 and 1310°F.
(C) Heat treatment above 1310°F.

(A) — Heat Treatment Below 1000°F.

17. The behavior of the metal was quite similar when heat treated at temperatures ranging from 500 to 1000°F. Therefore the results at one of these temperatures can be considered as being typical of this range. Table 3* gives the results of the tests at 1000°F. and these results have been graphically plotted in Fig. 1. The time in hours at temperature has been plotted against the percentage change in physical properties and combined carbon. Since no two samples of iron in the as-cast condition are absolutely identical in physical properties, it was found necessary to compute and plot the physical properties as well as the combined carbon in terms of per cent of the original in the as-cast condition in order to make the test results strictly comparable. The only exception to this procedure was in the case of the Brinell hardness, which in all instances has been plotted as the actual hardness number.

18. When the material is heat treated for as long as 96 hr.

* Refer to appendix where tables of observed data are given.

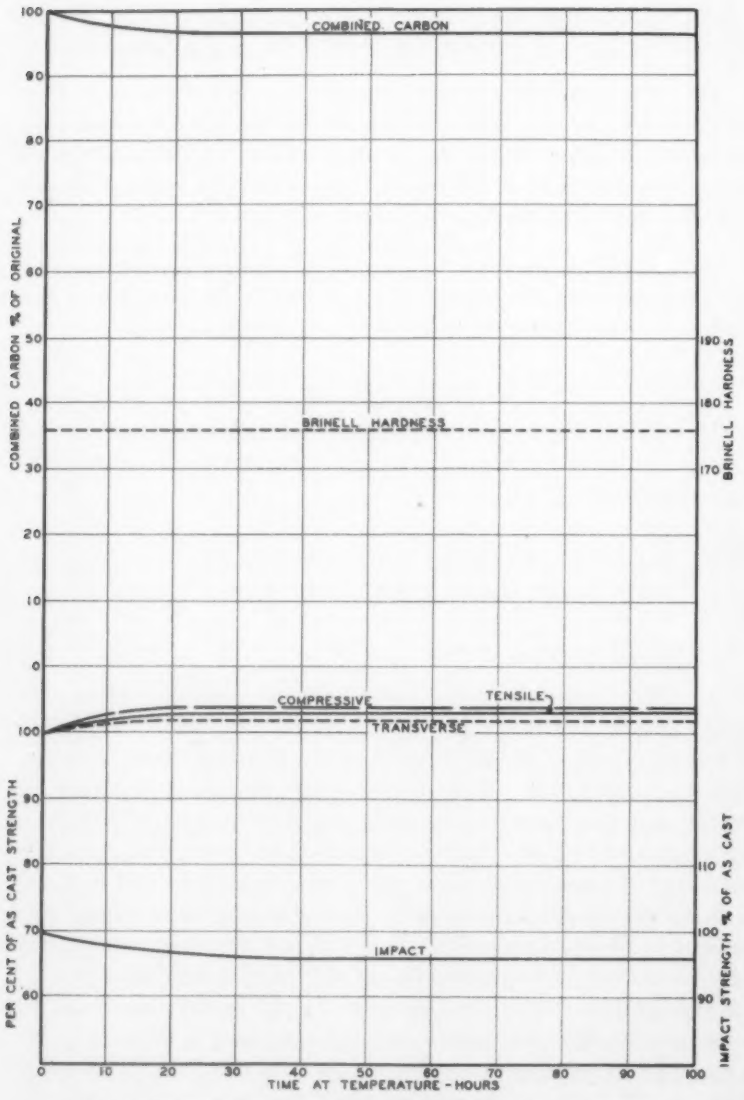


FIG. 1—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1000°F.

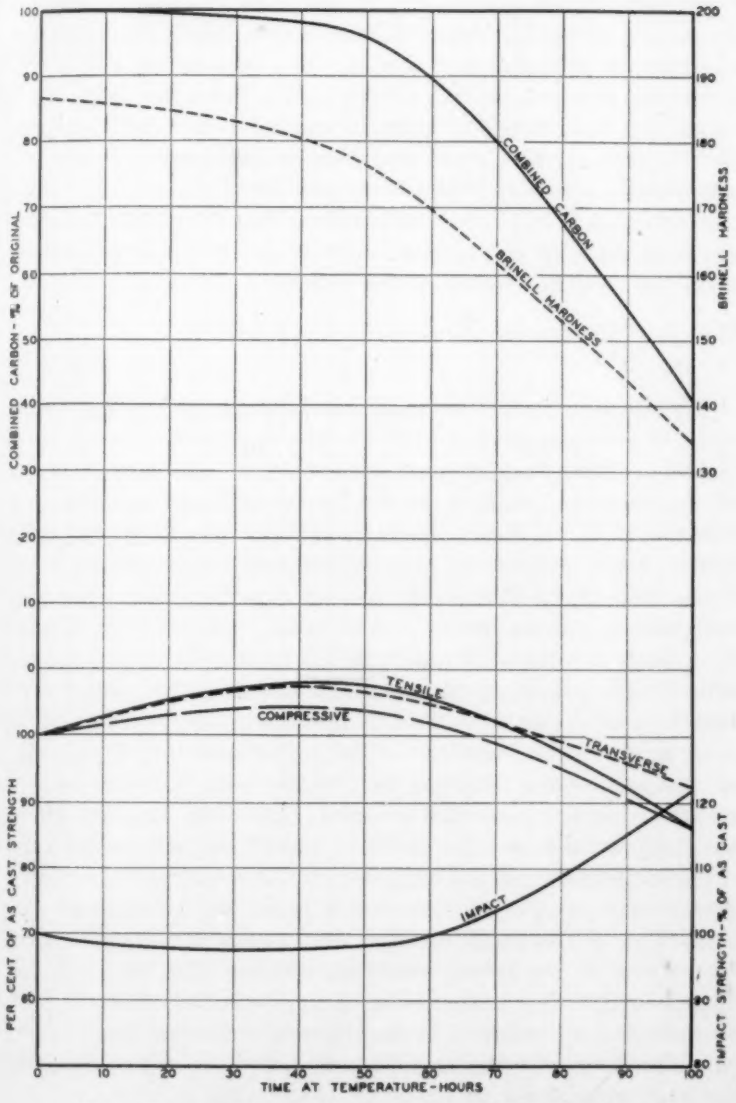


FIG. 2—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1100°F.

at any temperature up to 1000°F. the combined carbon and all of the important physical properties of the material are substantially unaffected. It should be noted that this behavior is in direct agreement with the dictates of the iron-carbon equilibrium diagram as is discussed in a subsequent section. At a temperature of 1000°F. the material is well below the critical range. It has not been transformed into the austenitic state to enable it to dissolve carbon from the graphite nor is the temperature high enough to cause a decomposition of the cementite in the pearlite of the matrix. Under these conditions it is only logical to assume that the physical properties of the material will remain constant due to the fact that the matrix has been unaffected by the treatment.

(B) — Heat Treatment Between 1000 and 1310°F.

(a) 1100°F.

19. Similar tests were made on bars which had been heat treated at a temperature of 1100°F. The results of these tests are recorded in Table 4 and plotted in Fig. 2. The data shows that the iron improves in physical properties when heat treated for a maximum of 48 hr. due no doubt to relief of stress. In excess of this heat treating time the physical properties and the combined carbon diminish. Although the combined carbon very gradually decomposes up to a maximum time of 48 hr., it begins to decompose very rapidly in excess of this time limit. Similarly, the Brinell hardness decreases at an increasing rate after 48 hr. The tensile strength, after 48 hr. at 1100°F., increases to 108 per cent of its as-cast strength, the transverse strength increases to 107 per cent, and the compressive strength to 104 per cent, but the impact strength remains practically constant. After 96 hr., the above mentioned values have fallen to 89, 95 and 88 per cent respectively, but the impact strength has increased to 120 per cent. The residual combined carbon clearly explains this behavior. After 48 hr. the combined carbon was still 0.76 per cent as against approximately 0.85 per cent in the as-cast condition, whereas after 96 hr. it had dropped to 0.38 per cent, a loss of approximately 50 per cent. The reduction in combined carbon likewise evidences itself in the Brinell hardness, it being only 70 per cent of the as-cast hardness after heat treating for 96 hr.

20. Thus, at 1100°F. there develops a definite tendency to decompose the cementite of the pearlitic matrix of the iron. The products of this decomposition are temper carbon and ferrite. Fer-

rite is low in tensile and compressive strength but high in ductility. As a consequence, the tensile and compressive strength are lowered and the impact strength increased with an increase in the ferrite content of the matrix. Transverse strength is known to be markedly influenced by the quantity of graphite or temper carbon in the metal, and hence, with an increase in this component the transverse strength diminishes. It is very important to understand that at temperatures below the critical range of the metal any reactions which take place are inevitably detrimental to most of the physical properties without any opposing reaction tending to improve them. Therefore, gray iron must not be subjected to temperatures just below the critical range of the metal for sustained periods of time.

(b) 1200°F.

21. The data for heat treatment at 1200°F. given in Table 5 may be considered as being extremely reliable for they represent the average result of four separate groups of tests. When the values are plotted graphically as shown in Fig. 3, a pronounced change in the behavior of the material evidences itself as compared to the results obtained by heat treating at temperatures lower than 1200°F. When the iron is heat treated at this temperature all of the physical properties, with the exception of impact strength, show a gradual but definite decline up to a maximum heat treating period of 24 hr. The losses then become very rapid to approximately 48 hr., after which they again assume a gradual decline to 96 hr. After being held for 96 hr. at 1200°F. the iron has retained but 63 per cent of its as-cast tensile strength, 69 per cent of its compressive strength and 72 per cent of its as-cast transverse strength. This rapid and pronounced reduction in these physical properties is due to the rapid decomposition of the pearlitic carbide evidenced by the extremely low residual combined carbon content. At this temperature the pearlitic carbide is extremely unstable and decomposes very rapidly without the presence of any opposing force which tends to recarburize the matrix. Below the critical range the iron is still in the alpha state and possesses a maximum solubility for carbon of 0.035 per cent as compared with a solubility ranging from 0.85 per cent to a maximum of 1.7 per cent carbon in the gamma state.

22. The only physical property which was not substantially impaired by heat treatment at 1200°F. was the impact strength

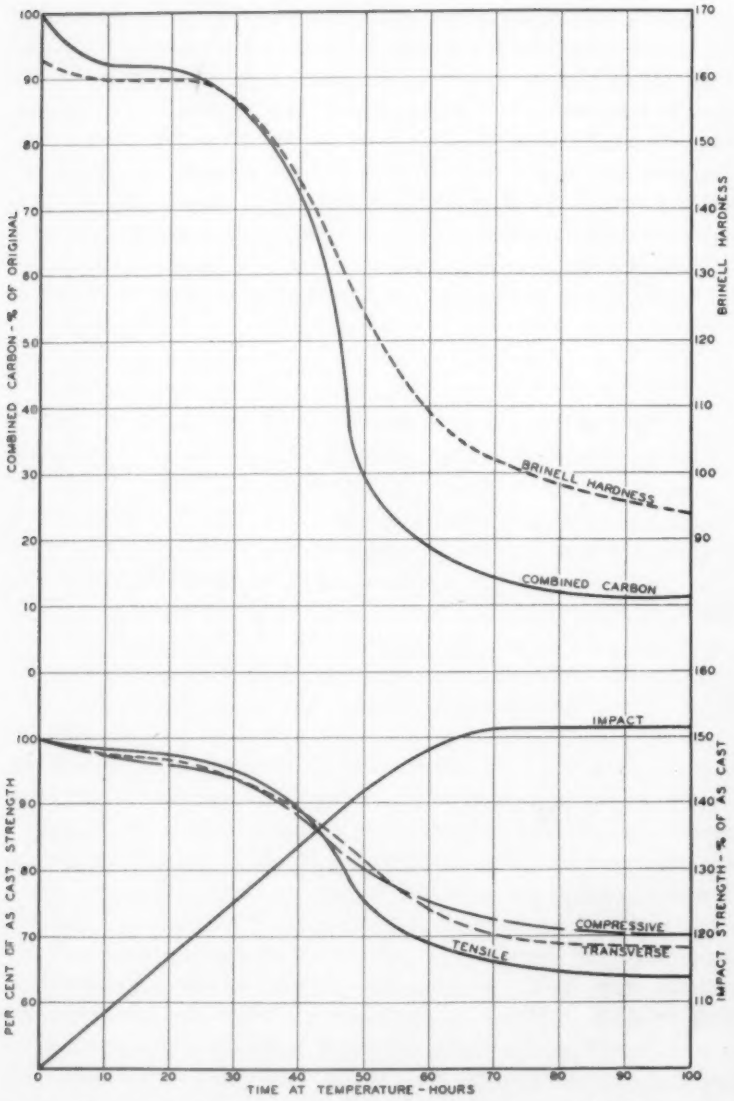


FIG. 3—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1200°F.

which gradually increased to approximately 150 per cent after 70 hr. and then apparently showed no further increase even after 96 hr. at this temperature. This behavior is also directly attributable to the decomposition of combined carbon. By referring to the curve showing the relation between combined carbon and time at temperature it will be noted that after 70 hr. the residual combined carbon has been decreased to 13 per cent of the original concentration in the as-cast state, but undergoes practically no change with additional heat treatment at 1200°F. The impact strength has been enhanced by the reduction in combined carbon. The ferrite which has replaced the pearlite in the matrix of the iron is much softer and more ductile, and imparts this property to the metal, thereby increasing the impact strength. Were it not for the fact that the metal loses in excess of 30 per cent of its tensile, transverse and compressive strength, it might be desirable to heat treat at this temperature so as to obtain this high impact strength.

23. Again referring to Fig. 3, it can be seen that the various physical property curves cross each other after about 44 hr. at temperature. Consequently, if the material is held for only 44 hr. at 1200°F. the metal will possess approximately 87 per cent of its original as-cast tensile, transverse and compressive strengths, and still will have acquired an impact strength which is about 135 per cent of its as-cast strength. The material in this condition will have a Brinell hardness of about 130 as compared with 160 in the as-cast state and a residual combined carbon which will be about 67 per cent of its original value. This concisely shows the value to be derived from complete test data. It affords an opportunity to arrive at the best physical properties obtainable under a given set of heat treating conditions. Because of the peculiar behavior of the metal when heat treated at this temperature, this heat treating cycle is not to be recommended for general use, however, due to the extremely close control of time and temperature essential to obtain the best results.

(c) 1310°F.

24. A series of test specimens heat treated at a temperature of 1310°F. (peak of the critical range on cooling) resulted in the physical properties recorded in Table 6. On heating, the peak of the critical range is approximately 1415°F., so that this group of test specimens were heated at a temperature just under the critical temperature. The behavior of the metal is almost the same as when

heat treated at 1200°F. but the changes are much more rapid. This can readily be seen by referring to Fig. 4 in which the results have been graphically recorded.

25. The combined carbon decomposes quite gradually until time at temperature exceeds 8 hr. and then the loss is extremely rapid reaching a minimum after 48 hr. From 48 to 96 hr. no change occurs, the metal having reached an equilibrium, possessing but about 17 per cent of its original combined carbon. The Brinell hardness curve follows very closely with the combined carbon curve, attaining a final Brinell hardness of but 105. Losses in tensile, transverse and compressive strength conform with the loss in combined carbon, reaching equilibrium values of approximately 64 per cent of the as-cast tensile strength, 73 per cent of the former compressive and 75 per cent of the original transverse strength after 48 hr. The impact strength gradually increases to a maximum value of 154 per cent after 72 hr. at temperature.

26. At this temperature the physical property curves cross one another after approximately 35 hr. However, the material then has a Brinell hardness of only 115 and a residual combined carbon content of but 22 per cent of that in the as-cast condition.

(C) — Heat Treatment Above 1310°F.

(a) 1400°F.

27. A group of slabs was heat treated at 1400°F. for various lengths of time ranging from 12 to 48 hr. The results obtained are recorded in Table 7, and the data graphically shown in Fig. 5. The impact strength shows a sharp rise to 138 per cent of the as-cast strength after 48 hr. at temperature. The other physical properties show a gradual decline reaching their lowest value between 24 to 30 hr. at temperature. After 48 hr. the tensile strength had dropped to 82 per cent, the compressive to 85 per cent, and the transverse strength to approximately 80 per cent of that in the as-cast condition.

28. Although the metal was still slightly below the critical range on heating (1415°F.) it is important to note that the residual combined carbon content reached a minimum of approximately 50 per cent of the original as-cast value after about 24 hr. at this temperature. The reason should be quite evident, for the metal was heated to a high enough temperature so that a portion of the pearlite had entered the critical range and thereby gained the

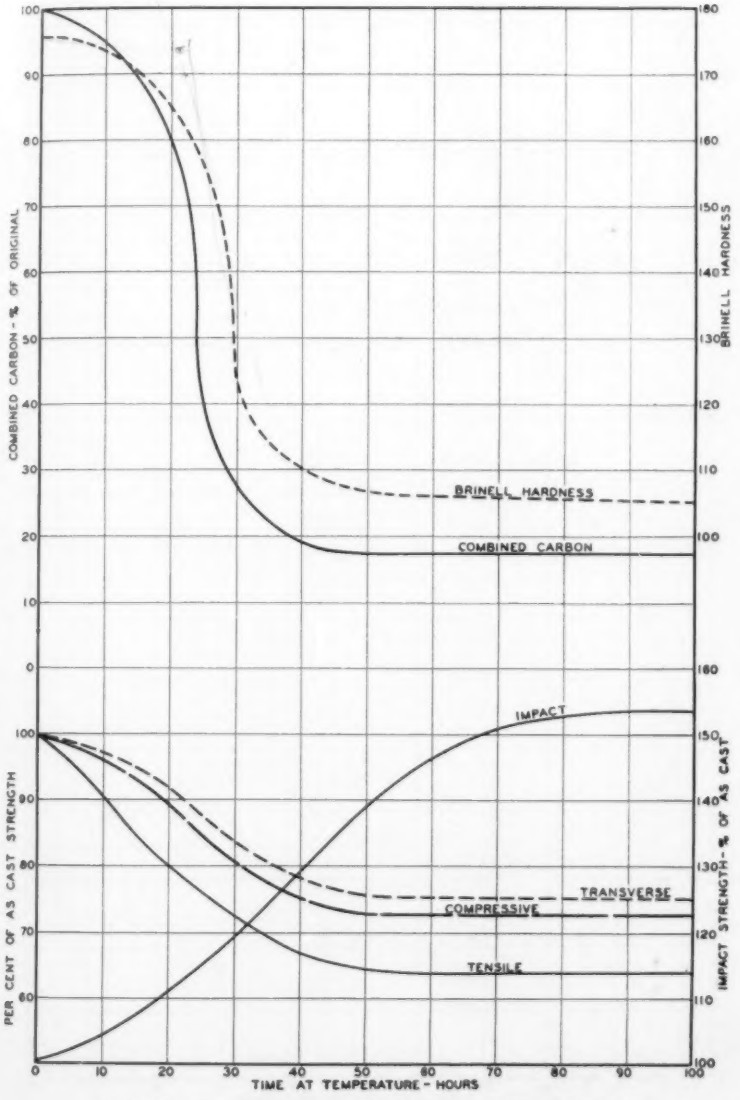


FIG. 4—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1310°F.

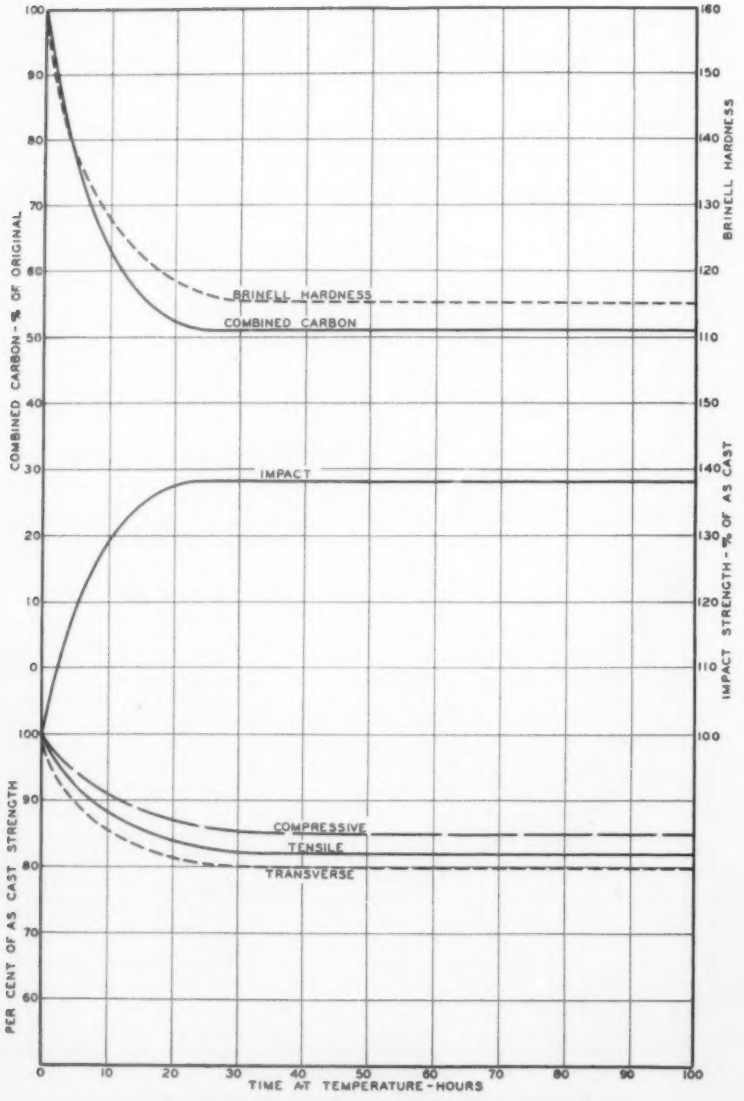


FIG. 5—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1400°F.

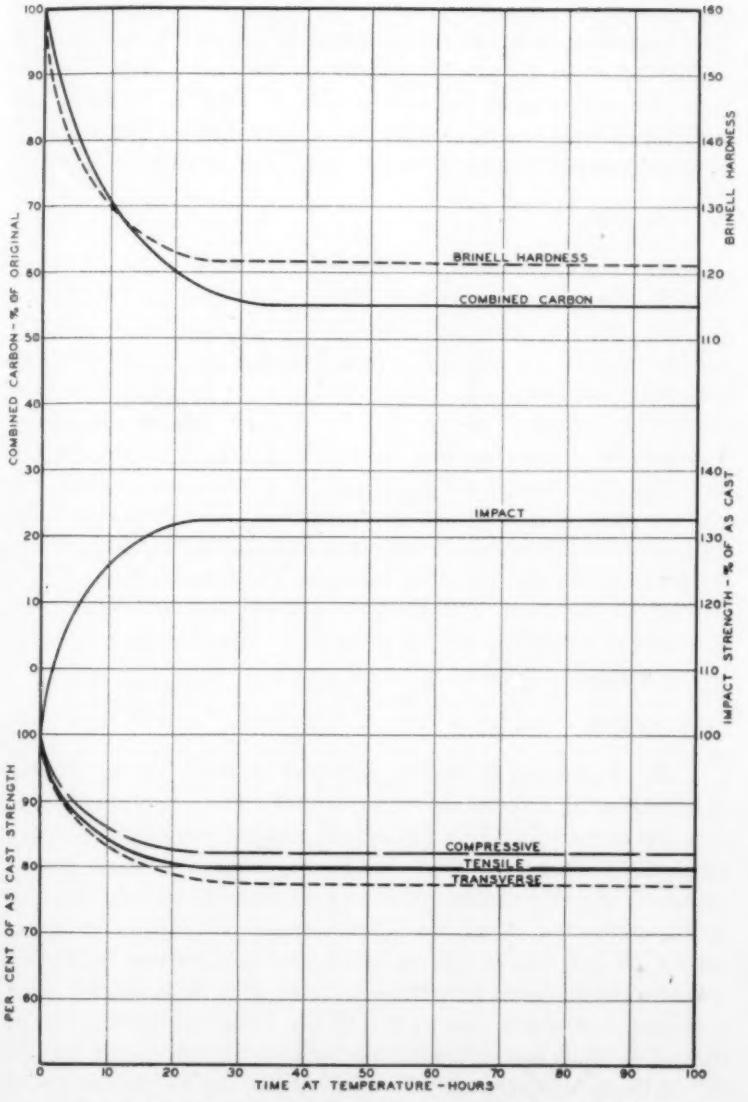


FIG. 6—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1500°F.

advantage of increased carbon solubility due to transforming into the austenitic state. Cast iron is a relatively heterogeneous alloy. Segregation of certain elements may cause a variation in the transformation range from one grain of pearlite to another. This readily accounts for the intermediate combined carbon decomposition. Some grains of pearlite persist at 1400°F. whereas others have transformed to austenite. The untransformed grains are subject to decomposition in the same manner as at 1200 to 1300°F.

(b) 1500°F.

29. The test results on specimens which were heat treated at 1500°F. for periods from 8 to 96 hr. are given in Table 8. The graphic representation of these results is illustrated in Fig. 6. Again, the loss in physical properties is very rapid, reaching a minimum and assuming equilibrium after 24 hr. The average values are as follows: 80 per cent of the original tensile strength, 82 per cent of the compressive, and 78 per cent of the transverse strength. The impact strength rose to a maximum slightly above 130 per cent, and then remained constant. This behavior is again a function of the combined carbon which showed a sharp reduction up to about 30 hr., but then assumes a constant value of 55 per cent of the original. The Brinell hardness similarly was reduced from 160 to an average of 122 after 24 hr., representing 76 per cent of the original hardness.

(c) 1600°F.

30. Following the same procedure, a group of test bars were heat treated at a temperature of 1600°F. for a minimum of 2 hr. to a maximum of 60 hr. The results are recorded in Table 9, and graphically shown in Fig. 7. The combined carbon decreased extremely rapidly at this temperature assuming its minimum value in approximately 12 hr. and then remaining constant at approximately 45 per cent of the original combined carbon in the metal as-cast. The physical properties likewise showed an abrupt change, assuming a constant value after 12 hr. Holding for this length of time at 1600°F. has reduced the tensile, compressive and transverse strengths to approximately 80 per cent of the as-cast values. The impact strength increased to about 123 per cent of the as-cast strength after 20 hr. at temperature. After 10 hr. the Brinell hardness was reduced to a minimum value of about 119, which is approximately 71 per cent of the original.

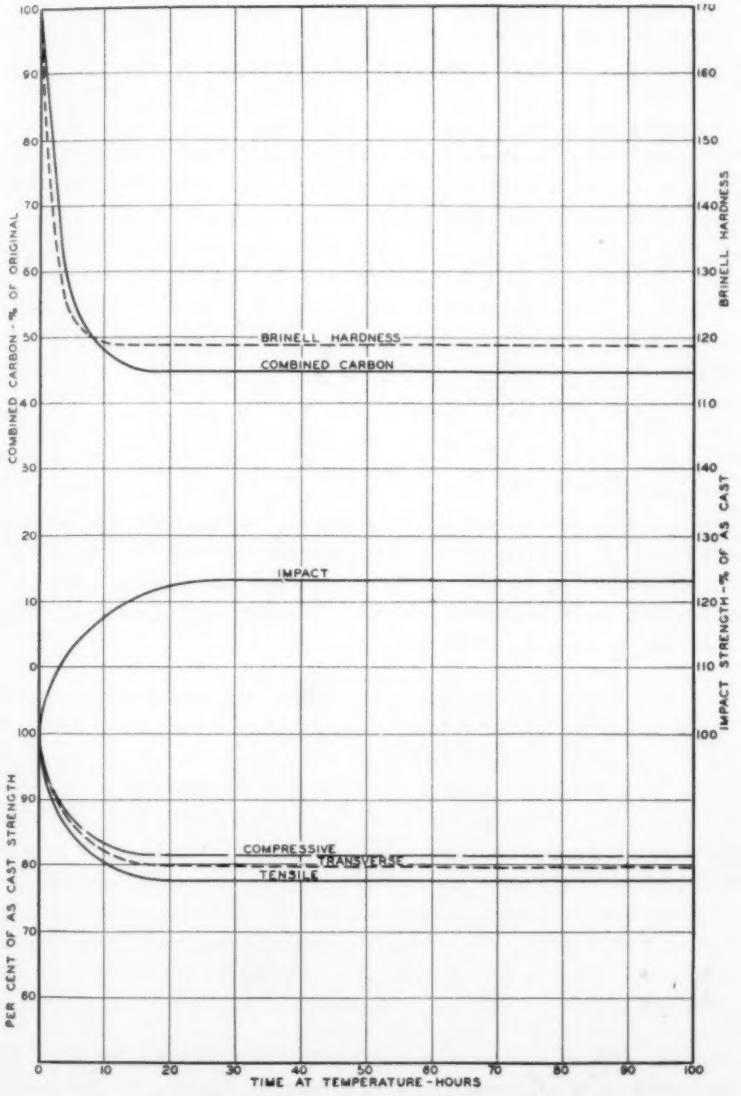


FIG. 7—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1600°F.

(d) 1700°F.

31. The test specimens which were heat treated at 1700°F. for varying lengths of time, ranging from 8 to 96 hr., show a similar behavior. The test results are shown in Table 10, and the values plotted in Fig. 8. The effect of heat treating even at this extremely high temperature is characteristic of the behavior of the metal when it is heat treated at any temperature above the critical range. The tensile, compressive and transverse strengths reached constant values of about 80 per cent of the original after 24 hr. The impact strength increased to approximately 127 per cent of the original after 48 hr. and the combined carbon was reduced to a minimum of 47 per cent of the original after 30 hr. After 20 hr. the Brinell hardness dropped to an average of 145, which was about 74 per cent of the as-cast value.

THE METALLURGY OF HEAT TREATMENT

32. It was felt that for the sake of simplicity it would be best to limit this discussion to the metastable (iron-iron carbide) equilibrium. Therefore, no mention has been made of the behavior according to the stable (iron-graphite) equilibrium or of variations in the carbon content of austenite induced by other elements in solution in austenite.

33. Wheel iron in common with all cast irons consists essentially of austenite and cementite just after solidification (Fig. 9). In the presence of sufficient silicon the behavior of the cementite between this temperature and ordinary temperatures is governed by the cooling rate of the casting. When cooling is rapid the primary cementite and the pro-eutectoid cementite, that precipitated between the eutectic and eutectoid temperature, are retained as such. At the eutectoid temperature the residual austenite transforms to pearlite. The final structure then is pearlite and cementite. This is the structural condition in the chilled portion of a wheel.

34. When cooling is moderately slow, the primary and pro-eutectoid cementite graphitize to a varying degree. In the mottled zone of the wheel, for example, the cooling rate is intermediate and such that only partial decomposition of the cementite occurs. However, in the plate and hub the cooling rate is such that the cementite is, for the most part, decomposed yielding graphite. At

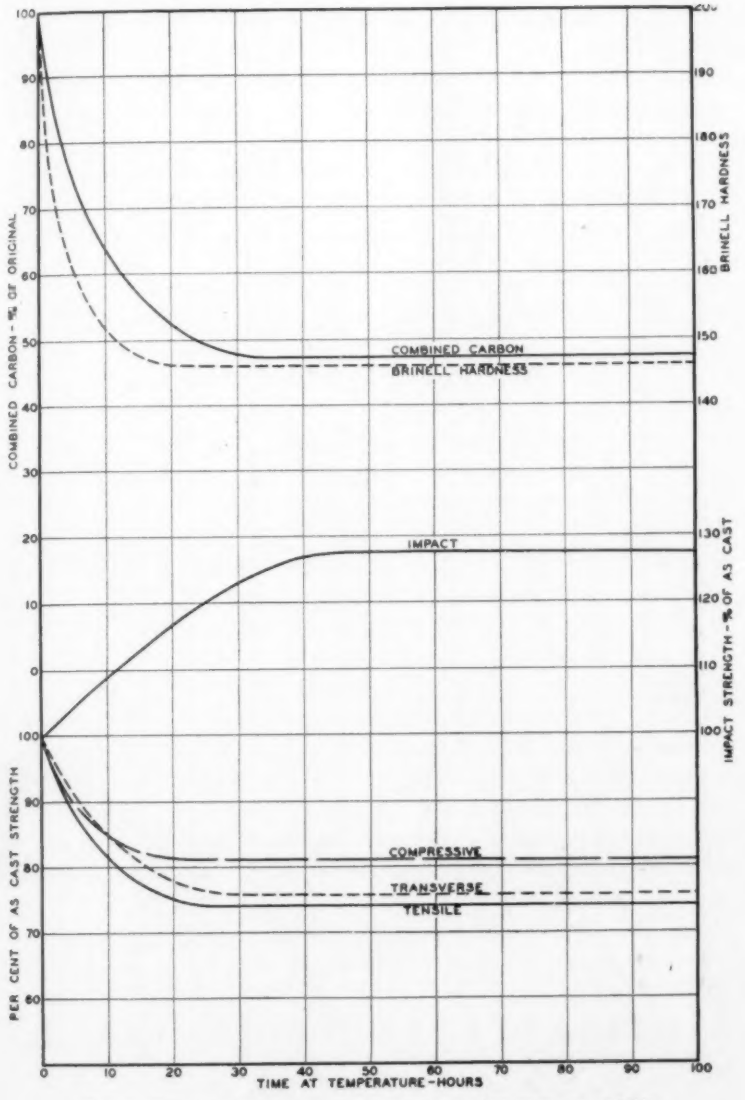


FIG. 8—PHYSICAL PROPERTIES OF WHEEL IRON ANNEALED AT 1700°F.

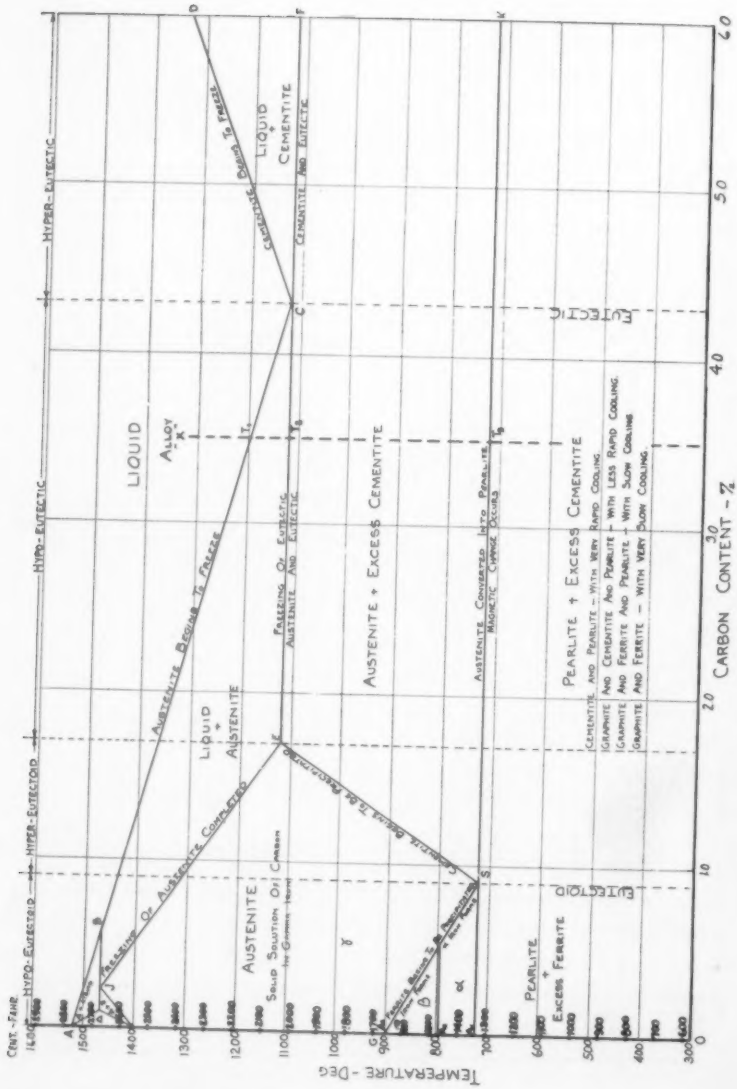


FIG. 9—IRON CARBON EQUILIBRIUM DIAGRAM.

the eutectoid temperature the residual austenite transforms to pearlite.

35. The carbide in the pearlite also would decompose if allowed sufficient time. The ultimate structure, then, would be one of ferrite and graphite. Normally, however, silicon is so balanced against the cooling time or section size in a particular casting that the pearlite remains unaffected and persists to room temperature. The final structure, is therefore, one of pearlite, graphite and perhaps some small areas of pro-eutectoid or primary cementite. This is the structure of the gray iron portions of a car wheel.

36. Upon reheating the changes take place in the reverse order. The cast structure will be unaffected at temperatures well below the critical range on heating even when the pieces are held at such temperatures for prolonged periods. Such treatment however will accomplish partial relief of residual casting stresses so that almost the full potential strength of the metal can be realized. As the temperature of reheating is raised to just below the critical range, say 1200 to 1300°F., the carbide in the pearlite exhibits a very strong tendency to decompose and will do so to a varying degree depending upon the time at temperature. At the same time any pro-eutectoid or primary carbides that may be present will also decompose. The same result could be obtained on cooling if the castings were held for a sufficient time at temperatures just below the critical range on cooling. In general, cementite decomposes most rapidly at temperatures in the vicinity of those at which it forms. Therefore, pearlitic carbide would be particularly unstable at temperatures below the critical range on cooling since it is here that the pearlite forms. When distribution of the graphite in a casting is good, the properties of the iron will depend almost entirely upon the structural conditions of the matrix. If the matrix is entirely pearlitic (0.6 to 0.8 per cent carbon retained in the combined form) the physical properties will be at a maximum. A lower concentration of combined carbon will result in a decrease in physical properties as has been clearly brought out in this investigation. The only property that shows an improvement is impact strength. This is due to the production of increasing amounts of ferrite in the matrix by heat treatment. Ferrite is tough and ductile and imparts ductility to the metal as a whole.

37. When castings are reheated to temperatures above the critical range on heating (1415°F. for chilled car wheel iron) the

pearlite present transforms into austenite. This form of iron has a marked increase in solvent power for carbon with increasing temperature between the eutectoid range and the eutectic range. Therefore at any temperature above the critical range on heating there will tend to be established an equilibrium between austenite and graphite, the carbon content of the austenite depending upon the temperature. Any primary or pro-eutectoid carbide areas that escaped decomposition during cooling of the casting now have sufficient time to decompose. Subsequent slow cooling through the critical range results in transformation of the austenite to pearlite and then decomposition of some of the pearlitic carbide to ferrite and graphite.

38. Thus the decrease in the amount of combined carbon upon heat treatment at such temperatures results from two causes, first, decomposition of any residual pro-eutectoid and primary carbide areas and secondly, decomposition of a portion of the carbide in the pearlite. It is interesting to note that the data shows a minimum limiting value of 0.4 to 0.5 per cent combined carbon to persist. Evidently, it is only possible to decompose a certain maximum amount of combined carbon by heat treatment at temperatures above the critical range. The final structure will therefore consist of pearlite, ferrite and graphite. Since this treatment results in some residual combined carbon one would expect the physical properties to be intermediate between those in the as-cast condition and those resulting from prolonged reheating at temperatures of 1200 to 1300°F.

39. It has been shown that decomposition of the combined carbon is rapid and complete at 1200 to 1300°F. and consequently the physical properties are severely damaged. Should this occur it is still possible to partially restore the metal to its original strength providing it is reheated to temperatures above the critical range. By this process the amount of combined carbon will be partially recovered due to the increased solubility of austenite for carbon. The graphite will supply the needed carbon to again enrich the matrix. Quite naturally, a corresponding recovery of some of the losses sustained in physical properties due to heat treatment at the lower temperatures will result from a second heat treatment at temperatures above the critical range.

40. A consideration of the data obtained from this investigation clearly portrays the behavior of the metal when subjected to

elevated temperatures for various lengths of time, and definitely confirms the dictates of the iron-carbon equilibrium diagram.

HEAT TREATMENT VS. MICROSTRUCTURE

41. The occasion rarely arises in which a definite change in the physical behavior of a metal cannot be either justified or even predicted in advance by making a micro-examination of the particular metal under consideration. The strength of a complex alloy such as cast iron is dependent upon the relative presence and quantity of certain constituents whose individual characteristics have been quite well established. For example, a gray iron containing a very large proportion of ferrite in the matrix will have properties which are quite definitely associated with the properties of the ferrite. Such a cast iron would be relatively mediocre in tensile and compressive but high in impact strength, due to the comparatively moderate ultimate strength in both tension and compression of ferrite. However, the high ductility of the ferrite will impart considerable resistance to impact stresses.

42. Similarly, if the matrix of the iron is rich in either primary or pro-eutectoid cementite, the metal will possess a higher ultimate strength in tension and compression but will be low in resistance to impact stresses. Here again the behavior of the cast iron is primarily governed by the constituent which is in preponderance, namely, the cementite which has high compressive strength but low ductility. Quite naturally a wide variety of conditions exist between these two extremes, and the physical properties of the metal likewise will lie between these extremes.

43. The behavior of the metal when subjected to various heat treating temperatures used during the course of this investigation can readily be justified by the micro-examination of certain typical specimens. For this purpose three typical cases were selected. The first had been treated at a very low temperature and represents the structure of the iron in the as-cast condition or when heat treated for a maximum period of 96 hr. at temperatures up to 1000°F. The second, which had been heat treated just under the critical range, is characteristic of metal whose physical properties had been very detrimentally lowered. The third represents the microstructure of metal which had been heat treated at a temperature of 1700°F. and is typical of the effect of heat treatment at

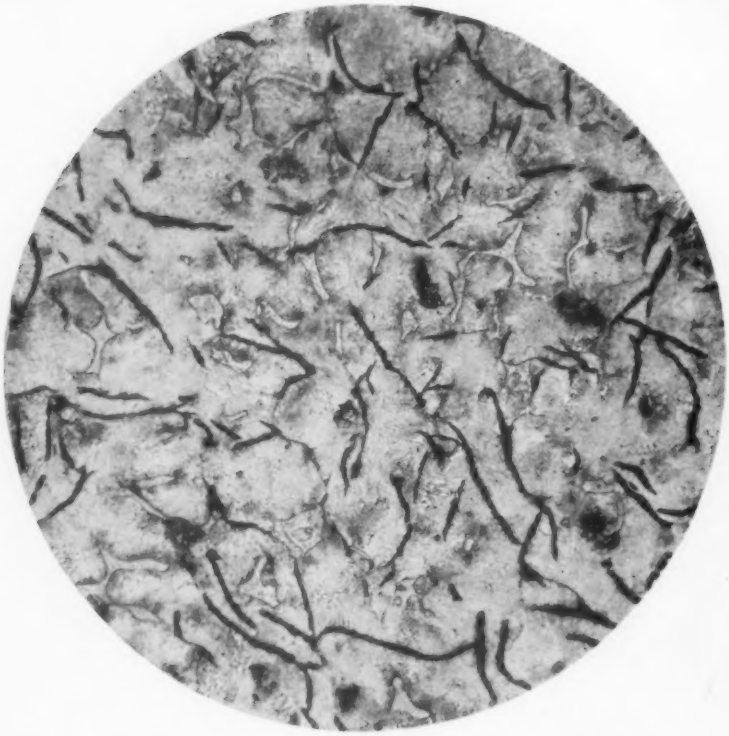


FIG. 10A—GRAY IRON HEAT TREATED FOR 96 HR. AT 500°F. (X100)

temperatures ranging from 1450 to 1700°F., this range being above the critical range of the iron.

44. Fig. 10A and 10B clearly show the structure at 100 and 1000 diameters of magnification of an iron which had been heat treated for 96 hr. at 500°F. This structure is typical of the plate metal in a car wheel in the as-cast condition or subsequently heat treated to any temperature less than 1000°F. The matrix consists of lamellar pearlite and some primary or pro-eutectoid cementite. The graphite distribution and formation is normal for this type of iron.

45. If the same metal is heat treated just under the critical range for 96 hr., as for example at 1200°F., the matrix is almost completely converted from lamellar pearlite (and some primary cementite) to ferrite as shown in Figs. 11A and 11B. Especially

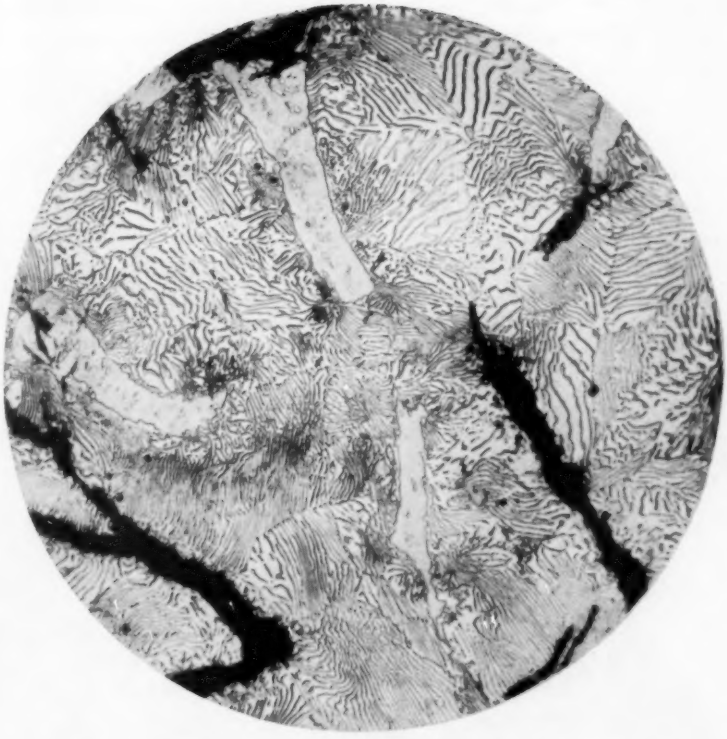


FIG. 10B—GRAY IRON HEAT TREATED FOR 96 HR. AT 500°F. (X1000)

at 1000 diameters, it is evident that the matrix is almost entirely ferritic. Minute pearlitic cementite lamellae still exist but their quantity is almost negligible. Small nodules of temper carbon are present resulting from the decomposition of the pearlite and primary cementite.

46. If the metal is heat treated at temperatures above the critical range the resulting structure is quite different. Figs. 12A and 12B show the structure of the iron after it had been subjected to heat treatment for 96 hr. at 1700°F. Instead of containing a ferritic matrix it contains at least 60 per cent pearlite. The primary or pro-eutectoid cementite has been completely decomposed. The important fact, however, is that the iron still possesses a matrix which is more than half pearlitic. This is further confirmed by chemical analysis for combined carbon. These analyses consistently showed a residual combined carbon content ranging from

0.35 to 0.45 per cent on all test bars which were heat treated for 96 hr. at temperatures above the critical range.

CONSOLIDATION OF FACTS

47. From this investigation certain specific behaviors and facts of importance have been derived and it is believed that a clearer conception of the importance of some of these points can be better understood by a consolidation. With this in mind the behavior of the material, when subjected to various heat treating temperatures, has been divided according to the particular physical property being considered.

Heat Treatment vs. Tensile Strength

48. Fig. 13 graphically shows the influence of time at temperature on the tensile strength of the metal at the various heat treat-

FIG. 11A—GRAY IRON HEAT TREATED FOR 96 HR. AT 1200°F. (X100)

ing temperatures studied. The tensile strength is shown in terms of per cent of the strength it possessed in the as-cast condition, considering the latter as being 100 per cent. Examination of this graph clearly shows three distinct types of behavior. At temperatures from 500 to 1000°F. the tensile strength remains practically constant, possibly showing a slight increase due to practically complete relief of internal casting stress. At temperatures just under the critical range, namely, from 1200 to 1310°F., the material sustains a very marked loss in tensile strength. The third group involves the heat treatment of the metal at temperatures above the critical range. At these temperatures the metal loses a comparatively moderate amount of its original tensile strength and then remains constant. Such a condition is highly desirable since a heat treatment of this kind lends itself well to production methods because it obviates the necessity for extremely close heat control.

FIG. 11B—GRAY IRON HEAT TREATED FOR 96 HR. AT 1200°F. (X1000)

Heat Treatment vs. Compressive Strength

49. In a similar way Fig. 14 shows the relationship between heat treating temperature and its influence on the compressive properties of the metal. The change in compressive strength by heat treatment at the various temperatures resolves itself into the same three groups, namely, that at temperatures up to 1000°F. practically no change takes place; at temperatures ranging from 1200 to 1310°F. a very severe loss in compressive strength is sustained, and at temperatures above the critical range, from 1450 to 1700°F., a comparatively moderate reduction in compressive strength results.

Heat Treatment vs. Transverse Strength

50. The transverse strength of the material after heat treatment at the various temperatures is shown in Fig. 15. It hardly

FIG. 12A—GRAY IRON HEAT TREATED FOR 96 HR. AT 1700°F. (X100)

appears necessary to again call attention to the fact that the transverse properties behave in a fashion identical with the tensile and compressive properties if the material is heat treated at similar temperatures.

Heat Treatment vs. Impact Strength

51. The impact strength of the material, as indicated in Fig. 16, shows no improvement if heat treatment is conducted at temperatures below 1000°F. Even at 1100°F. the impact strength of the metal does not improve perceptibly unless held at this temperature in excess of 60 hr. At temperatures ranging from 1200 to 1310°F. the impact strength shows a gradual increase attaining a maximum value after about 72 hr. No further improvement is shown for longer periods of time. At temperatures above the critical range, from 1450 to 1700°F., the impact strength im-

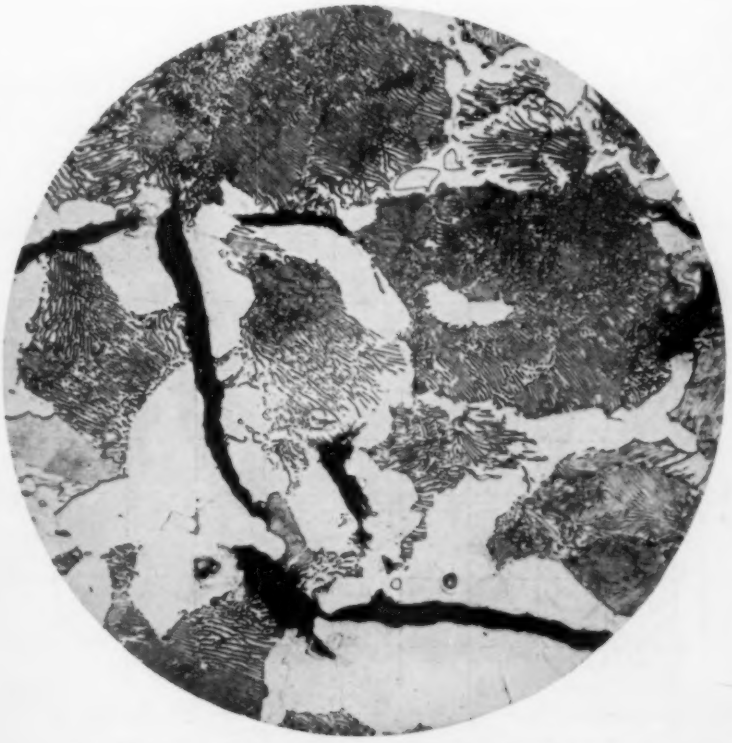


FIG. 12B—GRAY IRON HEAT TREATED FOR 96 HR. AT 1700°F. (X1000)

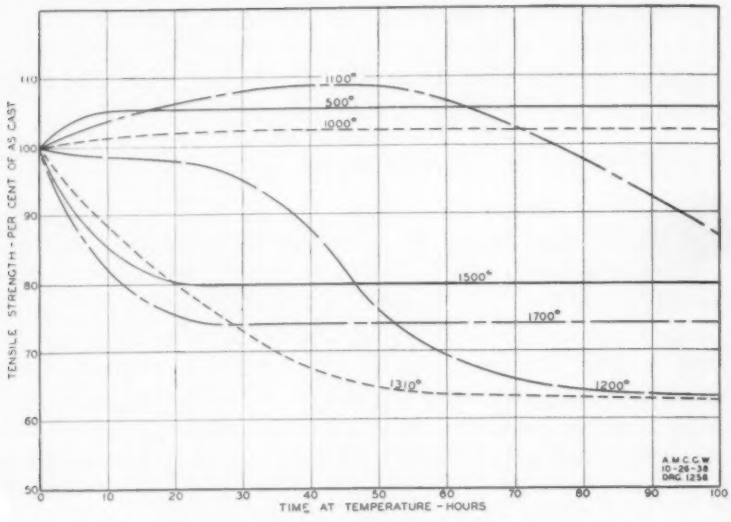


FIG. 13—TENSILE STRENGTH VS TIME AT TEMPERATURE CURVES.

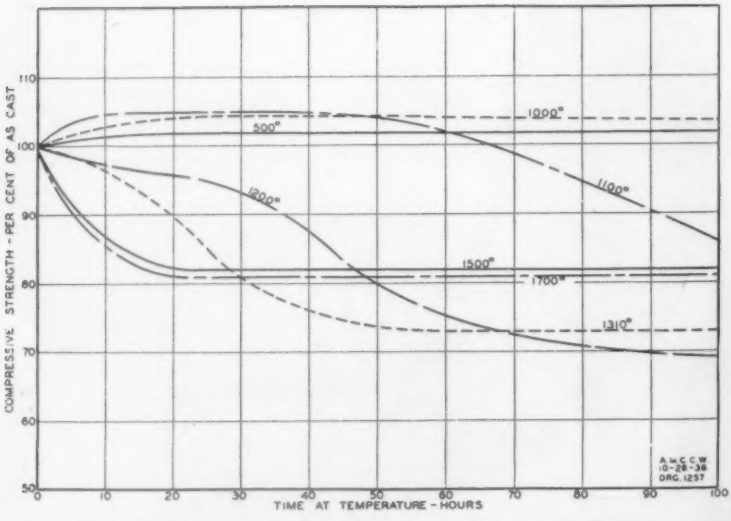


FIG. 14—COMPRESSIVE STRENGTH VS TIME AT TEMPERATURE CURVES.

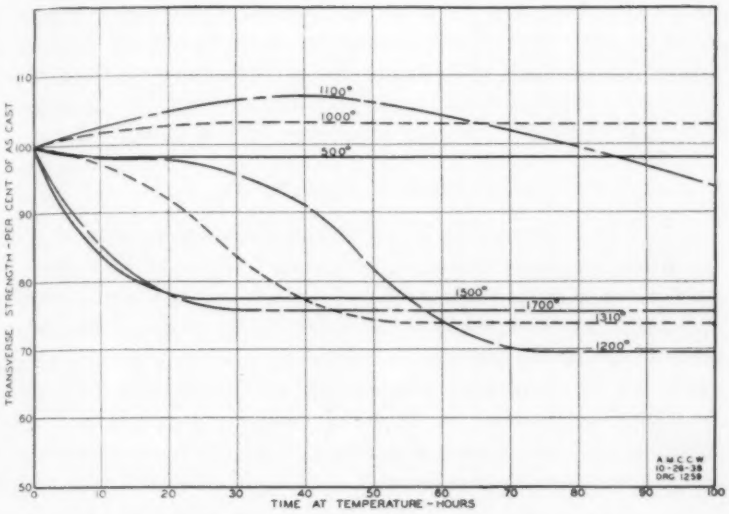


FIG. 15—TRANSVERSE STRENGTH VS TIME AT TEMPERATURE CURVES.

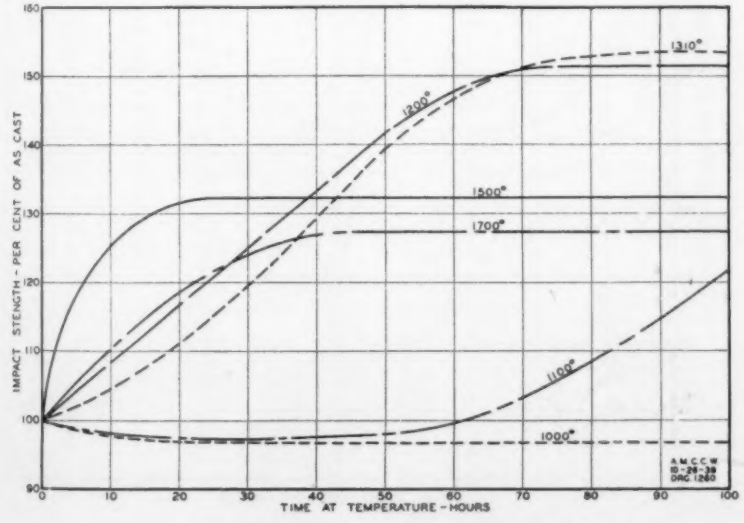


FIG. 16—IMPART STRENGTH VS TIME AT TEMPERATURE CURVES.

proves quite rapidly for so short a period of time at temperature as 24 hr., but no further increase is accomplished by holding for longer periods than this. Thus, the relation between time at temperature and impact strength is an inverse relation as compared to other important physical properties of the metal.

Combined Carbon vs. Time at Temperature

52. It is frequently highly advantageous to be able to predict the probable residual combined carbon content of the gray iron portions of a chilled wheel when heat treated at a given temperature. In order to make this convenient, Fig. 17 shows the relation between combined carbon in terms of per cent of the original when the metal is heat treated for various lengths of time up to a maximum period of 96 hr. at various temperatures. These curves again were almost a duplication of the curves showing the relation between heat treatment at various temperatures and the

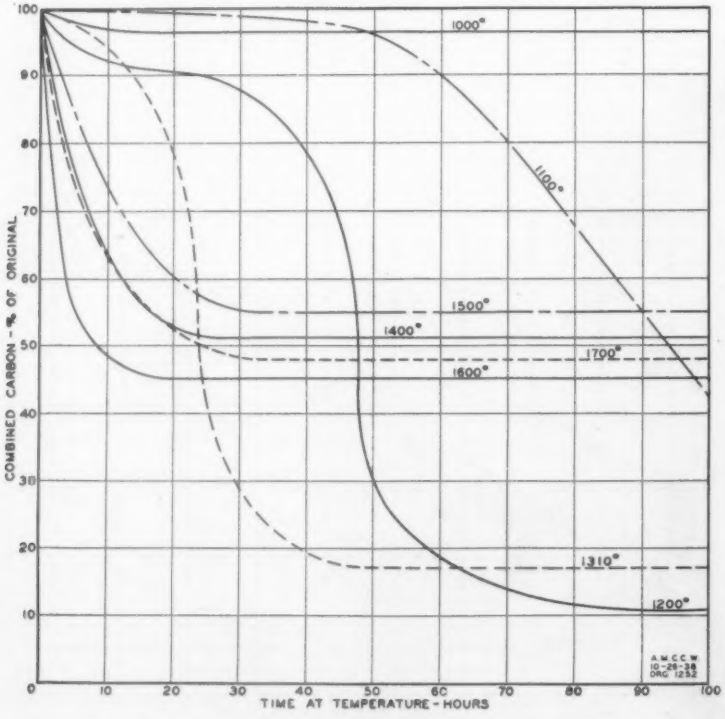


FIG. 17—COMBINED CARBON VS TIME AT TEMPERATURE CURVES.

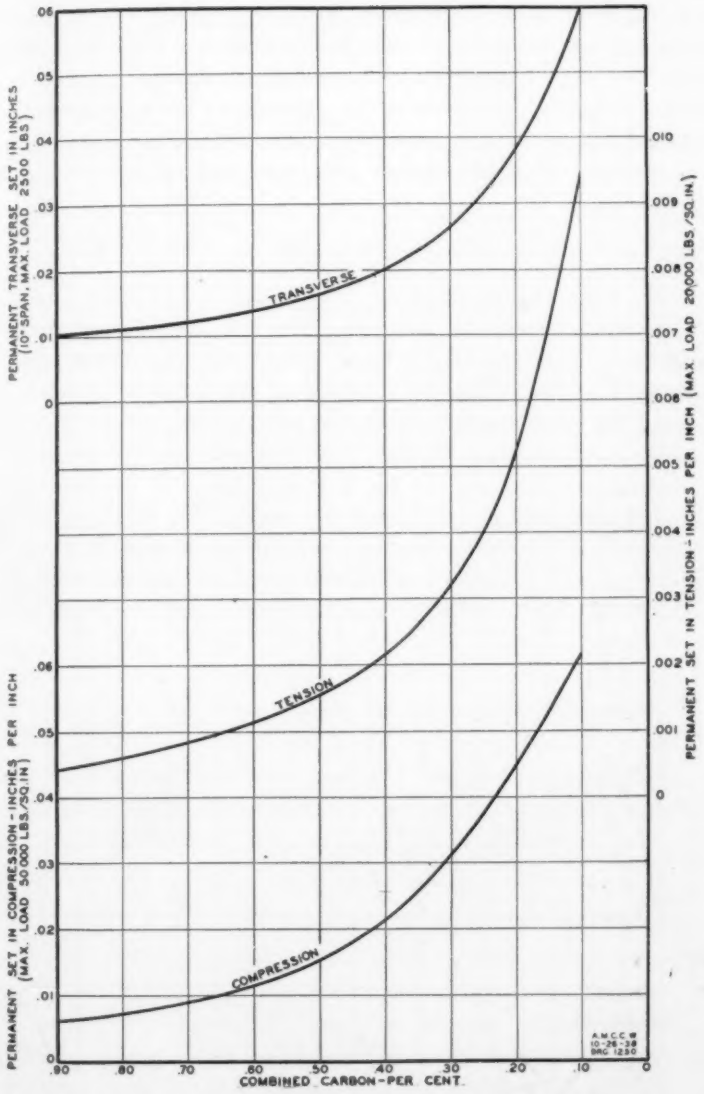

PERMANENT TRANSVERSE SET IN INCHES (10" SPAN, MAX. LOAD 2500 LBS.)
PERMANENT SET IN TENSION - INCHES PER INCH (MAX. LOAD 20,000 LBS./SQ.IN.)
PERMANENT SET IN COMPRESSION - INCHES PER INCH (MAX. LOAD 50,000 LBS./SQ.IN.)
TRANSVERSE
TENSION
COMPRESSION
COMBINED CARBON - PER CENT.
.90 .80 .70 .60 .50 .40 .30 .20 .10 0
0 .01 .02 .03 .04 .05 .06
0 .01 .02 .03 .04 .05 .06
0 .001 .002 .003 .004 .005 .006 .007 .008 .009 .010
A.M.C.C.W. 10-26-38 DRG. 1230

FIG. 18—PERMANENT SET OF WHEEL IRON VS COMBINED CARBON.

major physical properties. This data further emphasizes the very definite relation between physical properties and residual combined carbon. In all probability these same relations are also true for metal in the as-cast condition so that for a given combined carbon content, due either to low combined carbon content in the metal as-cast, or as a result of subsequent heat treatment, the physical properties are primarily a function of the combined carbon content regardless of how that combined carbon content was produced.

Permanent Set vs. Combined Carbon

53. The elastic property of the iron is the most important factor which governs the mounting pressure for a given tolerance, or difference in diameter between wheel seat and wheel bore. It is only the true elastic deformation of the material which serves to hold the wheel on the axle wheel seat. Any plastic deformation which may take place during mounting will slightly decrease the mounting pressure but in no way will assist in preventing the wheel from coming loose from the axle. The relation between combined carbon and permanent set, which in turn is a measure of the amount of plastic deformation which has taken place, is graphically shown in Fig. 18. On this graph is recorded the

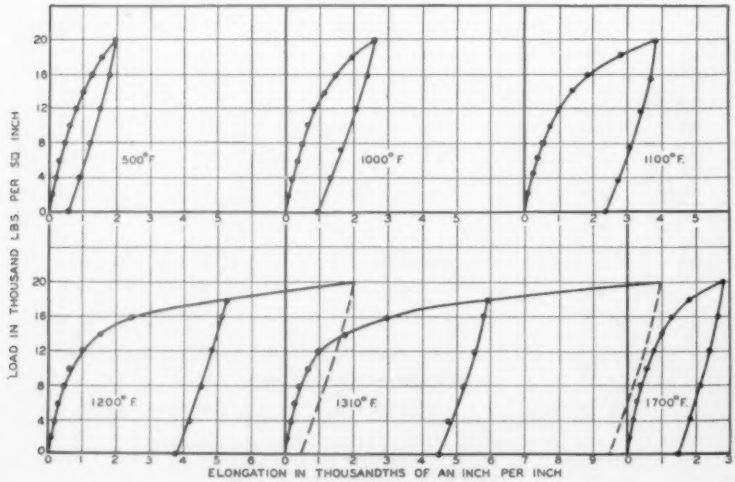


FIG. 19—TENSILE STRESS—STRAIN RELATIONS OF WHEEL IRON ANNEALED 56 HR. AT TEMPERATURES INDICATED.

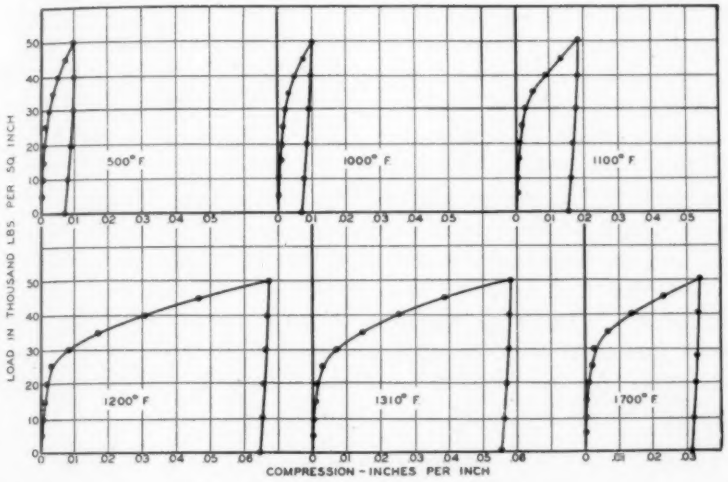


FIG. 20—COMPRESSIVE STRESS—STRAIN RELATIONS OF WHEEL IRON ANNEALED FOR 96 HR. AT TEMPERATURES INDICATED.

amount of plastic deformation when the metal is subjected to tensile, compressive and transverse loads. In each instance it becomes evident that the amount of plastic deformation (permanent set) is inversely proportional to quantity of combined carbon present, or in other words, as the combined carbon is decreased the amount of plastic deformation increases.

Elastic Behavior vs. Heat Treatment

54. A rather interesting determination of the elastic properties of the metal was conducted in connection with the investigation. Bars which had been heat treated to various temperatures were loaded to within 75 per cent of their ultimate strength in tension, compression and transverse, and the load was then removed. During the loading and subsequent relief of the load, extensometer measurements were made on the bar to determine the elastic behavior of the metal and the resulting permanent set after removal of the load. A typical record of such data is recorded in Table 11. The data so obtained has been graphically plotted in Figs. 19 to 22 inclusive. Since the ordinates represent the load in thousands of pounds and the abscissae deformation in inches, the area under each one of the curves represents work done in inch pounds. It is very evident that the elastic behavior (the work done during loading and subsequent relief of the load), is

a direct function of the residual combined carbon. At temperatures of 500°F. no loss in combined carbon was sustained and, as a consequence, the resilience was high, and the work done during loading and relief of load was comparatively small. At temperatures immediately under the critical range, such as 1200 to 1300°F., the work done becomes very large. When heat treatment is conducted at temperatures above the critical range, the work done again becomes relatively small. Similar behavior, although of varying magnitude, is typical of tensile, compressive and transverse loadings. To serve as a further comparison, Fig. 22 shows the elastic behavior or resilience of the metal in compression when heat treated at a temperature just below the critical range, namely 1310°F. As the time at this temperature increases, the elastic properties are impaired quite rapidly.

Physical Properties of Wheel Iron vs. Combined Carbon

55. Throughout this investigation the data has definitely and consistently directed attention to the fact that the physical properties are directly related to the combined carbon content of the metal. Consequently, Fig. 23 shows the relation between combined carbon and the tensile strength, tensile modulus of elasticity, compressive strength, compressive modulus of elasticity, impact strength and Brinell hardness. These curves are a consolidation

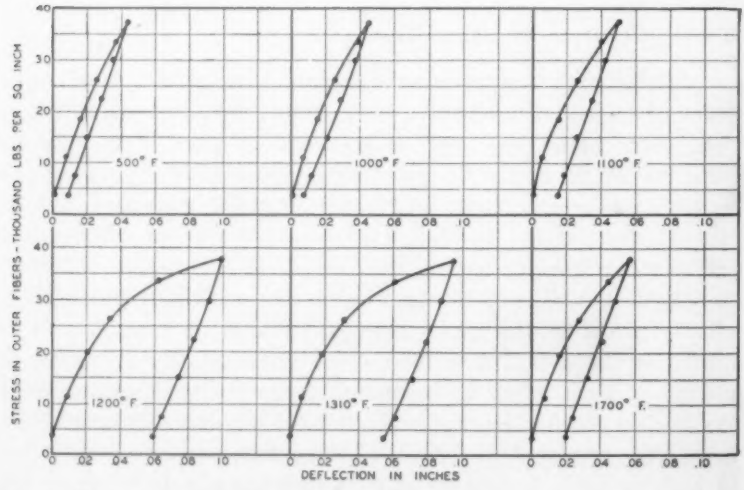


FIG. 21—TRANSVERSE STRESS—STRAIN RELATIONS OF WHEEL IRON ANNEALED FOR 96 HR. AT TEMPERATURES INDICATED.

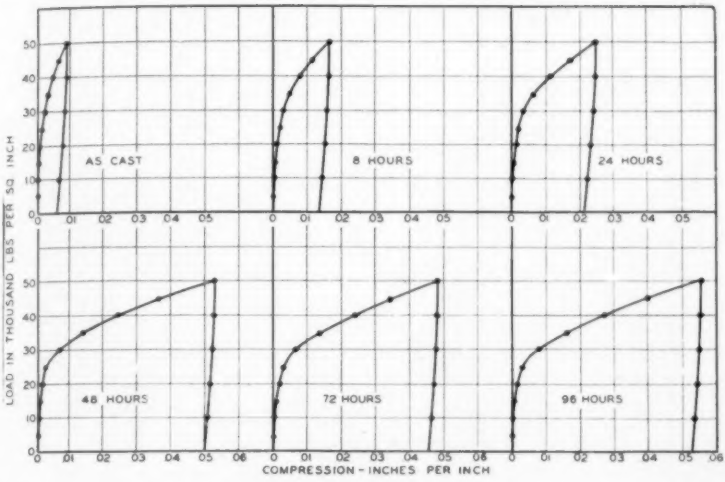


FIG. 22—COMPRESSIVE STRESS—STRAIN RELATIONS OF WHEEL IRON ANNEALED AT 1810°F. FOR HOURS INDICATED.

of all the data acquired from the testing of some 650 test specimens. The relations have been in excellent agreement and the data presented in this graph may be considered extremely reliable.

CONCLUSIONS

56. Heat treatment of a low silicon gray iron typical of the plate and hub of a chilled car wheel at various temperatures and for varying periods of time established the following points:

(1) No decomposition of combined carbon occurs when the iron is heat treated at temperatures up to and including 1000°F., for as long as 96 hr. Consequently the physical properties of the iron in the as-cast condition are unaffected.

(2) Heat treatment at temperatures just below the critical range causes substantial decomposition of the combined carbon, the degree of breakdown being dependent upon the temperature. Decomposition of the carbide is practically complete at temperatures ranging from 1200 to 1310°F., for example. Coincidentally most of the physical properties are lowered considerably. Impact strength is the only property that exhibits an increase with breakdown

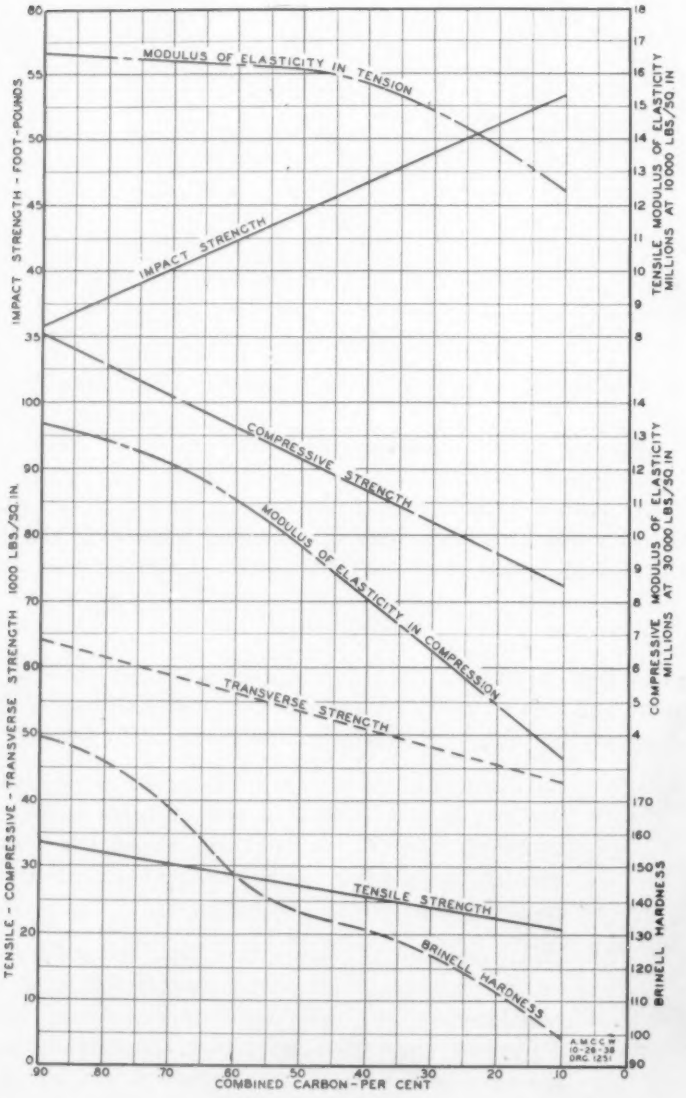

MODULUS OF ELASTICITY IN TENSION
IMPACT STRENGTH
COMPRESSIVE STRENGTH
MODULUS OF ELASTICITY IN COMPRESSION
TRANSVERSE STRENGTH
TENSILE STRENGTH
BRINELL HARDNESS
IMPACT STRENGTH - FOOT-POUNDS
TENSILE - COMPRESSIVE - TRANSVERSE STRENGTH 1000 LBS./SQ. IN.
TENSILE MODULUS OF ELASTICITY MILLIONS AT 10 000 LBS./SQ. IN.
COMPRESSIVE MODULUS OF ELASTICITY MILLIONS AT 30 000 LBS./SQ. IN.
BRINELL HARDNESS
COMBINED CARBON - PER CENT
A.M.C.C.W. 10-26-38 DRG. 1251

FIG. 23—PHYSICAL PROPERTIES OF WHEEL IRON VS COMBINED CARBON.

of combined carbon, due to the production of varying amounts of ferrite in the matrix of the iron.

(3) When heat treatment is carried out at temperatures above the critical range, 1500 to 1700°F., only a portion of the combined carbon is decomposed. In this temperature range an equilibrium value of approximately 0.4 per cent residual combined carbon is established. Under these circumstances the physical properties of the iron show an intermediate change.

(4) When distribution of the graphite is normal the physical properties of gray cast iron are dependent upon the structural condition of the matrix, and are best when the matrix is entirely pearlitic.

ACKNOWLEDGEMENT

57. The author is indebted to and wishes to take this opportunity to thank the Association of Manufacturers of Chilled Car Wheels for permission to publish this paper and for the assistance given by R. W. Lindsay, who read the manuscript and offered many helpful suggestions.

NOTE: Experimental data is found in the appendix on the next page.

Appendix

Tabulations of Observed Data

Table 3

TEST RESULTS — BARS HEAT TREATED AT 500° TO 1000°F. INCL.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
500-C	30,480	16,400,000	102,140	62,550	32.3	189.0	83.7	0.78
500-8	32,000	16,950,000	102,610	59,700	34.0	188.0	84.6	0.79
500-24	32,480	15,380,000	105,820	64,800	35.0	190.0	85.8	0.80
500-48	31,420	15,880,000	104,140	63,150	33.5	176.0	85.4	0.81
500-72	32,920	15,880,000	102,600	59,100	40.3	178.0	89.3	0.82
500-96	31,840	16,400,000	100,060	60,300	36.5	176.0	88.6	0.76
600-C	32,260	16,400,000	108,510	60,600	46.3	190.0	86.5	0.78
600-8	30,440	16,950,000	108,870	64,200	39.0	185.4	83.4	0.81
600-24	31,080	16,400,000	106,380	59,100	40.5	179.0	86.1	0.82
600-48	32,140	15,890,000	110,440	61,950	40.8	183.0	84.8	0.78
600-72	28,560	16,400,000	106,560	59,250	34.5	181.4	87.4	0.76
600-96	33,840	18,880,000	108,140	60,000	35.3	184.6	84.3	0.78
700-C	32,460	16,950,000	106,160	59,850	34.0	182.2	82.3	0.85
700-8	32,000	15,390,000	108,190	63,750	43.5	176.0	80.8	0.80
700-24	32,480	16,400,000	.....	63,600	39.5	177.0	84.2	0.78
700-48	33,340	15,880,000	105,550	60,750	39.5	170.0	82.4	0.81
700-72	31,140	15,880,000	110,050	63,450	37.5	176.0	83.6	0.85
700-96	31,060	17,540,000	106,670	59,700	40.5	176.0	85.7	0.82
800-C	31,120	15,880,000	104,380	61,350	38.0	184.6	86.0	0.78
800-8	31,160	17,530,000	105,040	65,250	39.0	191.0	86.5	0.79
800-24	33,440	16,950,000	106,700	62,100	48.3	194.0	85.5	0.79
800-48	31,180	16,950,000	104,060	58,050	42.0	192.0	81.5	0.81
800-72	32,580	16,400,000	105,400	63,000	36.5	181.4	84.5	0.78
800-96	32,540	16,400,000	104,740	58,350	33.5	188.0	81.5	0.79
900-C	31,100	15,390,000	102,140	54,000	38.3	179.8	82.3	0.75
900-8	31,680	16,400,000	100,930	60,900	36.0	173.2	84.4	0.77
900-24	32,400	16,400,000	100,720	58,650	41.5	169.4	84.8	0.74
900-48	31,700	17,540,000	101,140	55,950	38.5	172.4	84.1	0.77
900-72	30,780	16,400,000	102,780	60,300	39.0	179.0	86.0	0.76
900-96	31,700	16,400,000	102,030	58,200	42.0	189.0	83.8	0.78
1000-C	30,120	15,880,000	100,750	57,000	37.3	176.0	84.9	0.80
1000-8	28,480	16,400,000	107,660	60,750	39.0	181.4	85.9	0.77
1000-24	33,320	14,930,000	107,010	57,000	34.8	174.0	84.4	0.78
1000-48	30,840	16,400,000	101,780	58,800	36.0	173.2	83.9	0.76
1000-72	32,260	17,540,000	105,080	59,400	36.5	184.6	88.4	0.78
1000-96	29,160	14,100,000	103,750	54,000	33.5	175.0	85.0	0.77

Table 4

TEST RESULTS — BARS HEAT TREATED AT 1100°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1100-C	30,000	16,950,000	102,600	58,950	38.0	196.0	89.6	0.78
1100-8	31,220	15,400,000	107,160	61,950	31.5	181.4	89.4	0.78
1100-24	31,620	18,190,000	99,620	61,800	36.3	174.0	84.5	0.77
1100-48	33,660	15,400,000	107,180	57,150	37.0	179.8	86.4	0.76
1100-72	30,380	16,950,000	100,310	59,250	39.3	157.8	79.6	0.58
1100-96	26,620	13,700,000	90,210	55,950	45.3	158.8	75.7	0.38

Table 5

TEST RESULTS — BARS HEAT TREATED AT 1200°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1200-C	29,017	14,260,000	98,232	56,083	34.0	163.0	85.2	0.84
1200-8	28,483	13,243,000	96,173	54,500	36.0	159.7	82.0	0.78
1200-24	28,187	13,123,000	93,590	55,050	41.0	159.7	80.2	0.76
1200-48	22,777	11,700,000	80,017	46,950	47.5	128.3	67.2	0.29
1200-72	19,023	10,790,000	71,457	38,475	51.8	101.2	55.5	0.11
1200-96	18,380	12,050,000	67,607	40,150	51.1	93.9	50.4	0.09

Table 6

TEST RESULTS — BARS HEAT TREATED AT 1310°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1310-C	33,690	16,400,000	101,770	60,600	36.8	177.0	82.6	0.81
1310-8	30,440	16,400,000	100,890	57,750	38.5	168.8	81.8	0.78
1310-24	25,900	14,100,000	84,960	56,350	45.5	174.0	78.6	0.39
1310-48	21,740	12,660,000	79,280	46,050	50.5	106.6	60.3	0.14
1310-72	21,600	12,350,000	72,320	46,050	55.8	104.2	55.3	0.14
1310-96	20,280	14,100,000	72,650	41,550	56.5	114.8	57.4	0.14

Table 7

TEST RESULTS — BARS HEAT TREATED AT 1400°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1400-C	28,380	14,650,000	90,460	56,700	36.0	158.0	83.1	0.76
1400-12	23,530	12,990,000	81,485	45,450	49.0	123.0	64.2	0.41
1400-24	22,830	12,220,000	76,670	45,750	50.0	119.0	64.8	0.36
1400-36	24,500	12,540,000	79,185	45,450	52.0	115.0	60.3	0.38
1400-48	23,170	11,780,000	76,705	45,000	47.5	114.0	63.2	0.43

Table 8

TEST RESULTS — BARS HEAT TREATED AT 1500°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1500-C	29,580	14,300,000	100,850	59,700	33.5	159.8	84.1	0.84
1500-8	25,380	12,900,000	82,900	49,050	45.3	132.8	71.1	0.76
1500-24	22,660	11,600,000	84,000	48,000	43.3	122.2	66.7	0.50
1500-48	23,700	13,300,000	83,900	46,350	44.5	121.4	62.3	0.50
1500-72	23,180	12,800,000	79,930	44,550	45.3	114.8	62.9	0.43
1500-96	24,820	13,200,000	84,910	46,950	43.8	121.4	66.0	0.41

Table 9

TEST RESULTS — BARS HEAT TREATED AT 1600°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1600-C	29,030	13,890,000	95,275	55,350	37.0	170.0	82.0	0.76
B-1600-C	30,340	13,700,000	98,090	57,600	34.8	165.0	86.2	0.90
B-1600-2	27,170	13,700,000	90,575	54,550	42.5	138.0	72.8	0.66
B-1600-4	24,680	13,700,000	85,160	47,750	42.5	124.0	67.8	0.53
1600-6	23,900	11,370,000	81,365	47,400	41.0	122.0	61.9	0.37
1600-10	23,310	9,520,000	79,755	46,050	43.0	119.0	64.1	0.35
1600-14	23,870	10,420,000	78,035	47,100	42.0	122.0	64.2	0.37
1600-18	22,640	12,200,000	77,730	43,800	45.0	119.0	63.4	0.37
B-1600-36	22,530	11,620,000	78,940	44,250	41.8	117.0	62.7	0.33
B-1600-60	24,200	12,820,000	83,670	47,100	47.0	129.0	67.3	0.37

Table 10

TEST RESULTS — BARS HEAT TREATED AT 1700°F.

Sample	Ultimate Strength Lb. per Sq. In.	Modulus of Elasticity at 10,000 lb. per Sq. In.	Compressive Ultimate Strength Lb. per Sq. In.	Transverse Mod. of Rupture Lb. per Sq. In.	Impact Strength Ft. Lb.	Brinell Hardness	Rockwell Hardness	Comb. Carb. Per Cent
1700-C	36,400	16,950,000	112,780	63,900	41.0	195.0	86.8	0.81
1700-8	36,600	16,400,000	96,360	55,800	43.5	150.8	73.1	0.57
1700-24	30,240	15,880,000	93,770	55,350	48.5	144.2	70.9	0.45
1700-48	27,860	15,390,000	92,000	48,450	64.0	148.4	68.5	0.36
1700-72	25,140	16,400,000	95,000	51,450	52.5	144.2	60.6	0.36
1700-96	26,880	17,550,000	84,000	45,000	52.0	117.2	67.9	0.36

Table 11

TYPICAL STRESS — STRAIN DATA

1310°F. SERIES — ANNEALED 96 HOURS

Tension				Compression			Transverse			
Load in Lb.	Load Lb. per Sq. In.	Ext. Dial Read.	Elong. In. per In.	Load Lb. per Sq. In.	Ext. Dial Read.	Comp. In. per In.	Load Lb. in	Stress in Outer Fibers Lb. Per Sq. In.	Ext. Dial Read.	Defl. in In.
0	0	0	0	0	0	0	250	3750	0	0
1000	2000	0.3	.00006	5000	1.2	.00024	750	11250	8.5	.0085
2000	4000	1.0	.00020	10000	2.7	.00055	1250	18750	18.5	.0185
3000	6000	1.8	.00037	15000	4.7	.00095	1750	26250	31.5	.0315
4000	8000	2.6	.00053	20000	7.8	.00158	2250	33750	58.0	.0580
5000	10000	3.6	.00073	25000	14.5	.00294	2500	37500	83.5	.0835
6000	12000	5.1	.00103	30000	37.2	.00755	2000	30000	75.3	.0753
7000	14000	7.6	.00154	35000	82.2	.01669	1500	22500	69.0	.0690
8000	16000	12.2	.00248	40000	133.8	.02713	1000	15000	59.7	.0597
9000	18000	23.7	.00481	45000	196.3	.03988	500	7500	49.9	.0499
10000	20000	47.0	.00955	50000	275.0	.05583	250	3750	45.5	.0455
8000	16000	46.3	.00940	40000	273.7	.05550				
6000	12000	45.0	.00914	30000	271.0	.05500				
4000	8000	43.3	.00880	20000	267.9	.05435				
2000	4000	41.3	.00838	10000	264.7	.05370				
0	0	39.1	.00794	0	260.5	.05288				

(Table 11 continued on next page)

Table 11 *Continued*

Ultimate Strength	*Ultimate Strength*	*Ultimate Strength*			
10670 21340	68570	3030	45450	227.0	.2270

Impact Test	*Rockwell "B"*	*Brinell (3000 KG)*	
			Load
55.0 Ft.-Lb.	59.5		
60.0 Ft.-Lb.	54.5		
	52.0	5.90 5.80	5.85
115.0	54.5		
Average Impact	60.0	5.90 5.80	5.85
Strength = 57.5 Ft.-Lb.	58.5		
		5.90 5.80	5.85
	339.0		17.58
	Average Rockwell "B" Hardness = 56.5	Ave. =	5.86
		Brinell No.	100.6

Combined Carbon 0.20 per cent

DISCUSSION

Presiding: L. H. RUDESILL, Griffin Wheel Co., Chicago, Ill.

ALFRED BOYLES[1] (*Written Discussion*): The author is to be congratulated on the thoroughness of his investigation and the clearness with which it has been presented. One of the most interesting things brought out is the effect of time at temperature on the combined carbon in iron annealed above the eutectoid temperature.

According to the iron-carbon diagram, Fig. 9, a specimen heated to 1700°F. should consist of austenite containing about 1.25 per cent carbon in solid solution. The longer the time of holding at 1700°F. the nearer this condition of equilibrium should be realized. Upon cooling, the austenite should lose carbon down to the eutectoid composition and transform into pearlite, part of which may subsequently graphitize to give free ferrite.

Fig. 17 indicates that the amount of free ferrite formed is a function of the time of holding at 1700°F. in samples cooled at the same rate. Specimens annealed at 1400°F., 1500°F. and 1600°F. show the same trend. These curves suggest some change in the matrix of the iron which has reached completion after about 30 hr. at temperature inasmuch as longer annealing periods cause no further increase in the amount of free ferrite formed during cooling. While the present paper is not concerned primarily with theory, I should like to ask the author to give some comment on this point.

R. S. MACPHERRAN[2] (*Written Discussion*): It is a great pleasure to read an article in which the preparation has been so thorough as in this article on heat treatment of car wheels by Mr. Massari. The lower silicons in these irons render the action of heat a little slower and perhaps increase the temperature of change. The higher silicon irons are sometimes annealed a little easier. Otherwise, everything the writer says about chill iron can be applied to gray irons. We find then an annealing of much over 1000°F. usually results in a softer, weaker iron. Whenever heat is applied to castings in which strength and hardness

[1] Metallurgist, Battelle Memorial Institute, Columbus, O.
[2] Chief Chemist, Allis-Chalmers Mfg. Co., Milwaukee, Wis.

are essential, both the time and temperature should be carefully considered. The effect of prolongation of heating at 800-1100°F. on gray irons was described in a paper [3] presented before this Association by Rexford Krueger and the writer.

MR. MASSARI: The discussion prepared and presented by Mr. Boyles is greatly appreciated. The point referred to in the paper is of primary interest from a theoretical aspect. It is quite true that, according to the iron-carbon equilibrium diagram, at a temperature of 1700°F., austenite should contain approximately 1.25 per cent carbon in solid solution. Unless the sample is cooled extremely rapidly by quenching in order to retain the concentration existing at this temperature, the carbon concentration will follow the line *SE* on the diagram, the surplus iron carbide being precipitated until the eutectoid composition and temperature have been reached. At this point, the matrix will contain approximately 0.85 per cent carbon in solid solution if the alloy is one of iron and carbon only. It is quite probable that in a material as complex as cast iron, other elements present tend to reduce the eutectoid composition by translating the iron-carbon equilibrium diagram to the left. The exact location of this point is questionable and quite naturally varies considerably in various cast irons depending upon the concentration of other elements such as silicon, manganese, sulphur and phosphorus.

The gray iron utilized in this investigation contains some excess pro-eutectoid cementite. At temperatures above the critical range, this cementite is decomposed. Chemical analysis of the combined carbon content of the iron in the as-cast condition unquestionably includes the eutectoid carbon plus the carbon content in the pro-eutectoid cementite. Precisely speaking, these areas which are quite evident in Fig. 10-*A* are actually an iron-carbon-phosphorus solid solution rather than pure cementite. Nevertheless, they unquestionably are rich in carbon and materially contribute to the combined carbon of the matrix as determined by chemical analysis.

As Mr. Boyles observes, and as shown in Fig. 17, it appears that the combined carbon concentration in the matrix quite rapidly diminishes at temperatures above the critical and after heating for 30 hr. or longer, the combined carbon content of the matrix assumes an equilibrium concentration. In all likelihood, this reduction in the combined carbon content of the matrix is a function of two factors:

First, the decomposition of the iron-carbon rich constituent, which, for sake of simplicity because of its high carbon content, will be considered as pro-eutectoid cementite, and

Secondly, the decomposition of the eutectoid as the iron is cooling through the temperature range beginning at the eutectoid temperature and ending at a temperature of 1000°F.

The first of these reactions is a progressive, one taking place

[3] MacPherran, R. S., and Krueger, R. H., "Effects on Cast Iron of Prolonged Heating at 800-1100°F.," Trans., A. F. A., Vol. 38, 1930, pp. 826-855.

relatively rapidly and apparently completing itself during the first 30 hr., during which the metal is subjected to temperatures above the critical range. This condition is evidenced by the comparatively rapid reduction in combined carbon during the first few hours at temperature. The second reaction progresses more slowly and only takes place at subcritical temperatures.

After equilibrium has been established (30 hr. at temperatures above the eutectoid) the final combined carbon content is primarily a function of cooling rate through this subcritical temperature zone. Had these specimens been cooled more rapidly, a higher residual combined carbon content would have been retained. In order to definitely establish this fact, three specimens of the original material were heated for 36 hr. at a temperature of 1600°F. One of the specimens was then directly quenched in water to retain as nearly as possible the condition existing at this temperature. The second sample was cooled in air producing an intermediate cooling rate, and the last cooled in Sil-O-Cel, as all specimens were during the course of this investigation. Micro-examination of sections cut from these specimens revealed the following:

The metal which had been quenched in water from a temperature of 1600°F. had a matrix which was completely martensitic, without the presence of free ferrite, thus establishing the fact that rapid cooling through the subcritical temperature zone inhibited the decomposition of the eutectoid. The specimen which was air-cooled had a matrix which was almost completely sorbitic, but in addition, contained some free ferrite. The third sample which was cooled in Sil-O-Cel, and was cooled comparatively slowly through the temperature range previously referred to, had a structure typical of that shown in Fig. 12-*A* and *B*, namely that approximately 40 per cent of the matrix was ferrite. To further substantiate these findings, chemical analyses were run on the samples for total and graphitic carbon to make it possible for us to arrive at the combined carbon by difference. The results of these analyses are shown in the following table:

PER CENT COMBINED CARBON AS DETERMINED BY DIFFERENCE

	As-Cast	Quenched in Water from 1600	Air Cooled from 1600	Cooled in Sil-O-Cel from 1600
Combined Carbon (by difference)	0.82	0.85	0.72	0.50

The author also wishes to express his appreciation for the discussion submitted by Mr. MacPherran with whose statements he is in complete accord. Unquestionably, higher silicon irons will anneal much more quickly than those dealt with in this investigation, and it is quite likely that pronounced combined carbon decomposition and its accompanying weakening effect will take place in high silicon irons at a lower temperature and for shorter periods of heating.

M. G. CORSON[4] (*Written Discussion*): I am not going to criticize any of Mr. Massari's statements, for I find no doubtful points. I shall

[4] New York City.

only permit myself to indicate in a few words what further conclusions might be reached from Mr. Massari's data if a mathematical treatment were applied to them.

Let's take his sample 500-8 (page 40). We read: 32,000 lb. per sq. in. strength and 17,000,000 lb. per sq. in. elastic modulus. Combined carbon, about 0.79 per cent.

The matrix of this cast iron must possess a tensile strength of 105,000 lb. per sq. in., and a modulus of 30,000,000 lb. per sq. in. (steel). Allowing for a drop of 5000 lb. per sq. in. due to gas content, sulphur, phosphorus, etc. and dividing 32,000 by 100,000 we find 32 per cent. In other words, that is how much active area was left in the weakest section of this sample by the particular combination of the graphite flakes in it.

Divide 17,000,000 by 30,000,000 and you have 57 per cent. It means that the sample behaved, before it passed its proportionality limit, as if it were a bar of steel with a section 0.57 times as great.

But the average elastic section is not the true average section. The latter may be in all probability larger by 5 per cent. So we assume that the average section of that sample possessed about 62 per cent of active area.

On the other hand the best section available in a test bar of the usual length cannot possess more than 90 per cent active area. (A series of proper photomicrographs would permit computing the true maximum.) So with the help of these three figures we may plot an active area distribution curve as shown in Fig. 24. One must understand that this is only one of the many possible distribution curves, but this one is the most probable one on the account of eliminating any possible exaggeration. (It has to be drawn in such a manner that the areas between the lower and the upper part of the curve and the average area line are equal, and that there shall be no incongruous slopes.)

We repeat the same procedure with the sample 1600-10 (page 42). The combined carbon figure is 0.37 per cent. A matrix of this composition must have about 85,000 lb. per sq. in. strength and about 30,500,000 lb. per sq. in. elastic modulus. The actual figures are: 23,000 and 9,500,000 lb. per sq. in. The matrix may be again only 80,000 lb. per sq. in. strong on account of porosity and inclusions. In the manner outlined we find the minimum active area of 29 per cent and the elastic average of 32 per cent only. Add to it the 5 per cent and the average active area becomes about 37 per cent. So we trace a distribution curve as given in Fig. 25.

Compare those curves. The first is rather symmetrical. It shows that in all probability there is a uniform change of mechanical characteristics through the sample. The second is badly assymetrical. It shows that the sample contains many more sections of low than those of high

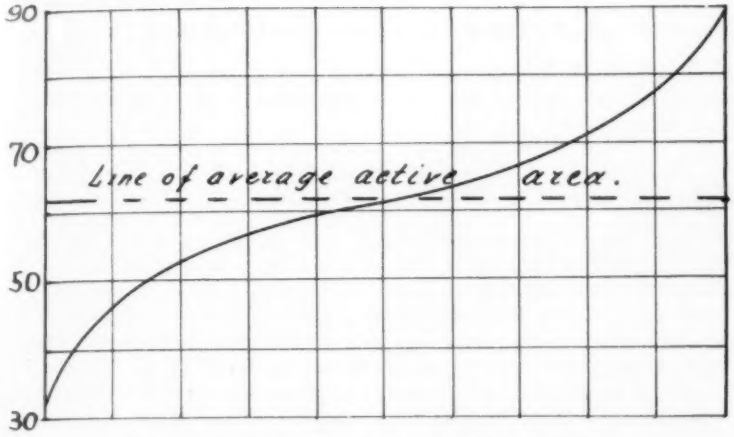


FIG. 24. DISTRIBUTION CURVE OF SAMPLE 500-8 (PAGE 40).

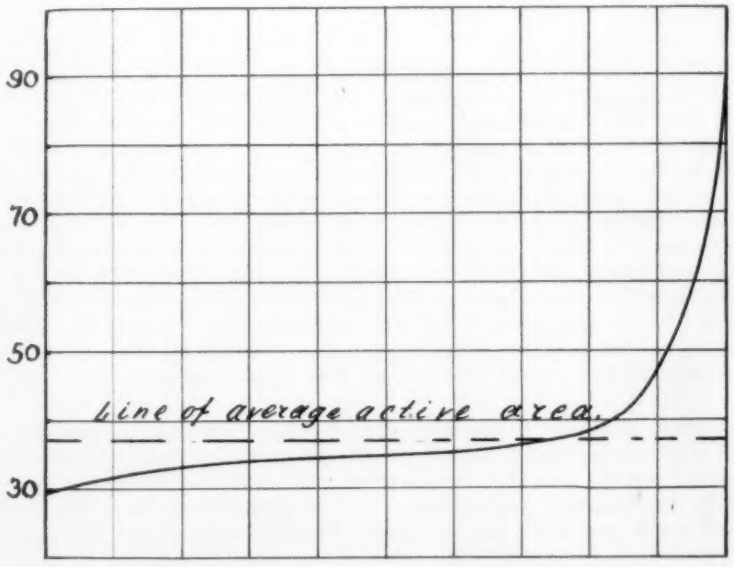


FIG. 25. DISTRIBUTION CURVE OF SAMPLE 1600-10 (PAGE 42).

strength. And that means that the particular heat treatment made the sample badly unreliable.

All this represents merely a hastily sketched picture. It could and should be greatly improved. It shows, however, that a mathematical

interpretation of the results obtained may be worth while. It can give us a deeper insight into the characteristics of cast iron.

I hope that Mr. Massari will find an opportunity to examine his figures and the data of his future researches in the light of the above considerations.

MR. MASSARI: I appreciate Mr. Corson's remarks and feel that this method of comparison should prove beneficial.

J. S. VANICK:[5] Mr. Massari has covered the ground so thoroughly that he has left very little room for argument and all I can do is to attempt to correlate some of this work with practical applications to foundrymen or practical applications which confront foundrymen. The first of these is with reference to the long-time heating at 1000°F., where apparently there is no loss in strength for about 100 hr. That should be interpreted as it stands and we should be careful not to assume that heating for longer periods will not cause some decrease in strength. Just how much, I think is a matter for some future work, but in our own work we find in heating for about 1200 hr., or perhaps a little longer, at 900°F. in steam, that the strength of an iron similar to this iron drops from 33,000 to 26,000 lb. per sq. in., or a loss of about 20 per cent.

The next point, which is extremely interesting to me and I am sure will be to machine tool builders in this Cincinnati area, is the relation to stress relief effects that a diagram such as the one illustrating the 1100°F. treatment on page 7 shows, where one finds this hump coming up in the strength after about 50 hr. of heating. Machine tool men are probably seriously considering the stress relief annealing of many of their castings and are working at temperatures somewhat under 1100°F. The period of 50 hr. at 1100°F. which the diagram shows necessary to get a hump in strength there might be too long for them to use but it obviously has some relation to the stresses in that casting use. It may be, and I believe it is quite likely, that something other than stress relief is going on, but I would prefer to have Mr. Massari add his explanation, because he has undoubtedly studied that very thoroughly.

MR. MASSARI: I appreciate Dr. Vanick's remarks particularly in view of our particular set-up. When working in an industrial research laboratory, we must do things that will deliver a monetary return. This investigation was prompted by the endeavor to learn the cause, in isolated instances, for extremely low residual combined carbon concentration in the hub of a car wheel. A car wheel is mounted on the axle by means of hydraulic pressure. The bore of the wheel is approximately 0.008 to 0.010 of an inch smaller than the diameter of the axle wheel seat and as the axle is forced into the wheel the elastic stresses that are established in the wheel serve to hold it on the axle. In other words, the total area of the wheel base multiplied by the coefficient of friction between steel and iron will give a definite measure of the elastic stress

[5] Metallurgist, International Nickel Co., New York City.

present, and a rather close approximation of the force with which the wheel is held on the axle.

As I determined during the investigation, the elastic return of the metal when the combined carbon content is reduced to a low value is comparatively small. The metal in this condition largely undergoes plastic deformation. Plastic deformation does not serve to hold the wheel on its axle; it is only the elastic stress since plastic deformation results in permanent set which in no way assists in holding the wheel on the axle. We have had isolated instances in which wheels having a very low combined carbon under 0.10 per cent have actually come loose from the axle, and, since the journal and journal box are an outboard bearing, the wheel actually moved to the center of the axle.

In an endeavor to determine the cause for such abnormally low combined carbon content, we undertook this investigation so that today I am confident we are in an excellent position to warm the wheel maker to avoid sub-critical temperature zones and if a wheel of this kind is encountered, we can definitely attribute the particular factor in the process of manufacture which was responsible for this undesirable condition.

I have no good answer for the point which Dr. Vanick has raised. If there are structural changes that have taken place, I am unaware of them. I am rather firmly of the opinion that the increase in strength is an outward evidence of internal stress relief.

The Foundry and Metallurgical Science

By M. G. Corson,* New York, N. Y.

Part I—Present Non-Ferrous Foundry Alloys, Their History and Development

Introduction

1. Details of foundry work and casting alloys have been discussed in scientific and technical literature for the past 50 years. However, the apex of objective foundry knowledge reached thus far consists of the material compiled in a few handbooks, a number of variable standards and a few rules setting forth the best foundry practice to use in given cases. In addition, we have a number of nebulous hopes and not a few prejudices which we are prone to accept as true knowledge.

2. On the other hand, no general philosophical treatment of the whole field has ever been attempted. Twenty years ago, when metallurgists were lacking in the foundry industry, this situation may have been justifiable. Today the situation is vastly different. More and more technical men go into the foundry game and a general review of the situation may be really worth while. So without expecting to give a hundred per cent true picture by his individual effort, the present author is attempting to take stock of what we possess in the foundry field in its relation to metallurgical science.

A Brief Historical Review

3. It is a thankless job to hunt for precise data to determine when and where a certain alloy or a method originated. Secrecy enshrouded most foundry developments for scores of years and the situation became so confused that a search for origins would call for high grade detective talent which the author does not claim to possess. So, of necessity, the chronological presentation may suffer in regard to its precision.

*Consulting Metallurgist.

Note: Presented Before Non-Ferrous Session, 43rd Annual A.F.A. Convention, Cincinnati, Ohio, May 16, 1939.

Tin Bronzes

4. *Historical.* The 10 per cent tin bronze alloy seems to be the main, if not the only, alloy which humanity developed at a very early date. Starting to use tin as an alloying element about 4500 years ago, man came to recognize the composition mentioned as the most convenient one not later than 3500 years ago. Since then, other variants have been found, such as the speculum metal of the ancients with its 32 per cent tin. The 20 per cent tin bronze used for the casting of bells was also known at least 2500 years ago. Nevertheless, the 10 per cent alloy remained the universal standard for thousands of years.

5. The reason was that, in addition to yielding reliable castings, this composition could be worked into many shapes with the aid of nephrite hammers and quartzite anvils, plus enough patience and a power of observation. For many centuries, bronze was used for axes and chisels with cast bodies and hard hammered edges, swords of cast and hammered strips, and brooches of bronze wire, some highly ornamental. All this was accomplished in the complete absence of rolls, wire-drawing benches and the like. This art of hammering tin bronze into intricate shapes constitutes the only really lost art.

6. So, while prehistoric bronze-smiths were active simultaneously in hundreds of places in Europe alone 3000 years ago, few foundrymen of today are even sure that a 10 per cent tin bronze can be cold worked successfully. The author has seen cases where small ingots of a 5 per cent tin bronze were subjected to experimental cold forging under steam hammers and the job given up in despair because the 3-in. square ingot became too hard to work further when reduced by a mere ⅛-in. in thickness!

7. In the thousands of years since the development of 10 per cent tin bronze by prehistoric smiths, ordinary brasses in their wrought form and gray iron for castings (to say nothing of wrought iron) came into use. Nevertheless, the art of casting bronze continued to occupy a place in mankind's economy, being transferred from father to son and never improving, but rather deteriorating. With the coming of the wrought iron age, the need

for making almost perfect bars of cast bronze to be worked into tools and weapons disappeared. Bronze became a material used mainly for bell and statue casting and there no one cared whether the casting was good inside as long as it showed no surface cracks.

8. New demands came when artillery was developed. True enough most of the small calibers were cast of gray iron or forged together from smaller pieces of wrought iron at an early date. However, the larger pieces were cast in bronze by the same people who developed the art of casting large bells. The old flame furnace used in the bronze foundry, was of a larger capacity than the primitive cupolas of the gray iron foundries.

9. Three improvements came much later. In the early part of the nineteenth century, the Austrian general Uchatzius developed his method of improving field guns by forcing a steel mandrel into the cast piece and then hammering the casting on the outside. This constituted practically a re-introduction of the old industry of cold hammered bronzes, but the superior technique of the new steel making industry cut short the life of this recovered art.

10. *Introduction of Phosphorus and Zinc.* Much later, in the 1870's, the art of improving bronzes by adding phosphorus or zinc, previous to the introduction of tin, were developed. The first was established in the industry of phosphor-bronze, the second in Admiralty gun metal. No guns were ever cast in this new 88-10-2 composition, but for some reason, it became the darling of all the navies in the world. In fact, that composition is still used for a large number of articles used on ships.

11. Phosphor bronze did not fare so well in the foundry field. The addition of phosphorus to metal that has to be cast is an extremely delicate operation. Very small quantities do not work any improvement; somewhat larger ones result in blackened castings with many pinholes. Perhaps the only fields where cast bronzes may contain a substantial amount of phosphorus are gear blanks and certain bearings. The first is the more important, while the second becomes even more limited.

12. The production of gears, gear blanks and similar castings, resulted in the firm establishment of alloys containing between 11 and 12 per cent tin in the foundry industry. Below 11 per cent, the metal is too soft; and above 12 per cent, it is too brittle, so the limits are rather close.

13. It was found later that a 14 per cent tin bronze, while completely lacking in ductility and having the mediocre strength of not over 35,000 lb. per sq. in., was hard enough to withstand abrasion and dense enough to remain tight under high gas pressures. Therefore, it has become a standard material for certain parts of compressors and the like. It is a strange but established fact, that sound castings are most likely to result with 14 rather than with 10 per cent tin or any other alloy. Stranger still, this high tin alloy rarely, if ever, gives birth to the *tin-sweat* which admiralty gun metal and the gear blank compositions quite frequently do.

14. *Red Brass.* When and where the red brass composition with its three fives—5 per cent tin, 5 per cent zinc, and 5 per cent lead—was first used, is a question. Obviously, a composition of this kind cannot be expected to show any outstanding characteristics for there is no magic connected with the "three fives." However, this composition was extremely easy to remember and the "man at the fire" who saved a little money and started a shop of his own, liked such simple formulae. Perhaps he started with the still easier formula: one ounce of tin, one of lead, one of zinc to every pound of copper—the so-called ounce metal. Later, when switching over to percentages in the sophomore stage of his existence, the practical foundryman took a fancy to the "three fives."

15. Since then, a huge amount of scrap, in this particular composition, has accumulated. While it is quite possible to use gear or admiralty gun metal scrap in compounding the "three fives" composition, going the other way is more than difficult. Again, brass scrap, and even some babbitt, may be allowed in the red brass charge with little if any harm.

16. So today, the "three fives" is quite an established product. It is cheaper than all others, and whenever no responsibility is attached to the use of a casting, the red brass in the States, the ounce metal in England, and the rotguss (red castings) in Germany, is the first alloy to be specified. Very rarely tested and still less frequently studied, its usefulness is sanctioned by its usage.

17. Oddly enough, while admiralty gun metal was long considered THE metal for responsible jobs, it gave plenty of trouble in its applications. The castings frequently were unsound, uniform results were difficult to secure and they did not become less so even

after a number of investigations were undertaken to study the influences of temperatures, molding sands, additions and impurities. It was found that lower temperatures are more likely to give better castings, that small amounts of lead improve machinability, that small amounts of nickel improve strength and that small amounts of antimony and arsenic are harmless. But still, a safe and sure way of securing castings that would look sound and be capable of withstanding high hydraulic or steam pressures, has not yet been found. Even the pyrometrical control and the purchase of ingot from the most reliable maker, does not guarantee results in the simple castings, to say nothing of very complicated ones.

18. For this reason, probably, experiments were started in the direction of so changing the mixture of the gun metal as to obtain better results, while still sticking to the essential type of the alloy. Whether a thorough series of such experiments was ever undertaken, the author does not know, but certainly it was never reported. Nevertheless, such varieties were "found" and we now have the steam bronze and the valve bronze, perhaps a few more special bronzes, the maximum variation running from 10 to 7 per cent in tin, from 1 to 4 per cent zinc, and perhaps up to 3 per cent lead.

19. *Highly Leaded Bronzes.* Literature gives no indication as to when highly leaded bronzes came into use. It is only logical to assume that the idea came as a corollary to the use of ounce metal. At any rate, it seems that bronzes containing 10 per cent tin and 10 per cent lead were in regular use as railroad bearings early in this century. Later came those containing 20 per cent lead with the same amount of tin, and because this composition is likely to contain surface pinholes, intermediate compositions, most of which contain a lower amount of tin, a reasonable change, also came into being.

20. The culmination points in high leaded bronzes are represented by two alloys: a 30 per cent lead, 5 per cent tin plastic bronze and another alloy containing no tin at all but as high as 40 per cent lead. These two alloys were the basis of strong competition and controversy some 30 years ago, but do not seem to be of much importance today.

21. Apart from industrial tin bronzes, we still possess the old bell metal and statuary bronze. The first varies considerably—

the monotonous harsh bells of locomotive being cast from a metal very different from that of the carillons. The metal of cymbals is probably also quite different. Still more variable is the metal of the statuary bronzes, in which the amounts of tin, zinc, and lead vary in the extreme. That is to say, the analytical data obtained by various authorities show a great variation. Whether the specialists in statuary castings long ago reached a conclusion as to which composition is best for a given job, we cannot say. At any rate, bells and statues form a "specialty" industry into which neither science nor empirical knowledge has yet taken even a brief peep.

Brasses

22. It is difficult to decide when and where common brasses began to be used for castings. We may be sure that no regular castings in brass were executed until Gerhardt Muntz developed his 60:40 metal in 1832. While Muntz was in the wrought brass business, the startling discovery that a brass might not be hot short, set some foundrymen to wondering. It was found that the same metal could yield decent castings and quite cheaply too.

23. Almost all later derivatives from Muntz metal were used, for a time at least, in the foundries. Therefore, an enumeration of these derivatives might safely apply to the range of the present discussion.

24. *Iron Bearing Brasses.* Perhaps the first one among these represented Muntz metal with varying (but small, always less than 1.5 per cent) additions of tin and lead. The first increases the strength somewhat; the second profoundly enhances machinability.

25. Later came the great series of false alarms—the iron bearing brasses, which were supposed by their inventors to combine the characteristics of brass with those of the elusive nebulous metal called "strong steel." Alloys containing a bevy of components, among which iron occupied its position to the extent of 0.5 to 2 per cent were recommended for castings, forging, etc. under different names. Lack of definite information among metallurgists, and especially among their customers, permitted their claim of a wonderful achievement when the cast metal possessed 70,000 lb. per sq. in. strength with 10 per cent elongation, although this is just a bit below the best value obtainable with a 42 per cent Muntz metal containing 1 per cent tin.

26. *Manganese Bronze.* Later came the really strong brasses, every one of which contained at least 1 per cent aluminum and some as high as 3.5 per cent of that metal. Usually, some manganese and a bit of iron also were present, but their influence, if any, never was proved definitely. Nevertheless, these high strength brasses became known as manganese bronzes. Their naming had some reasons, however, since a copper base alloy's name does not carry the idea of strength unless the term bronze is included in it. The name aluminum bronze was already attached to a different composition so the combination of the word manganese with bronze was hit on as both euphonious and imposing.

27. Manganese bronzes may possess, according to their aluminum content, a tensile strength ranging from 80,000 to 115,000 lb. per sq. in. (as cast). Obviously enough, a range of this width allows of subdivision into a number of types, and because the aluminum added is recovered to the extent of 95 per cent or more, it is easy enough to control the composition with good precision. However, three types are most regularly offered. One contains but little aluminum, the second just enough to retain from 20 to 40 per cent of the alpha constituent, and the third is a straight beta alloy, quite close to its limit of saturation with aluminum and zinc.

28. Probably the largest tonnage of sand cast manganese bronze belongs to the second class and the bulk of it goes into the manufacture of ship propellers. Here a strength of 80,000 to 90,000 lb. per sq. in. may be guaranteed, according to the amount of beta constituent present; *i.e.* the amount of aluminum added. The strongest alloy does not seem to be widely used and the weaker one is probably merely listed.

29. *Yellow Brass.* Finally, we may note yellow brass—an alloy containing about 25 per cent zinc plus 2 per cent tin and 3 per cent lead. This alloy, while stronger and cheaper than red brass and gun metal, very rarely is used and hardly ever subjected to testing. Probably its use is limited to those cases where some non-rusting, cheap, metal of a yellow color is needed to fill up a hole in space.

30. This situation is strange because all brasses can be cast easily whether their zinc content is 10 or 40 per cent, and with one per cent lead, they can be machined excellently. Their corrosion resistance is at least as good as that of any tin-base bronze and it is

difficult to see where they might prove inferior, except when working under load and friction. But the more successful alloy, red brass, is not much used in such circumstances anyway.

Aluminum Bronzes

31. Still stranger is the situation in the domain of aluminum bronzes. They were used for castings, tentatively at least, long ago while aluminum itself was merely an industrial curiosity. Later, when scientists began to work on copper-aluminum alloys, a subject beloved of scientific investigators, it was found that bronzes with up to 8 per cent aluminum were too plastic and those with 10 per cent too unreliable (self annealing). Then again, it appeared to be well established that aluminum bronzes are likely to contain films of aluminum oxide and to yield dirty, porous and leaky castings. So for many years, aluminum bronzes were looked upon as poison.

32. Later, the interest in aluminum bronzes was kindled anew, for one reason because it was discovered that an addition of 3 to 4 per cent iron to an alloy with 9.5 per cent aluminum, leads to strong reliable castings without any trace of self anneal. While nothing radical was done to cope with those "dangerous films" of aluminum oxide, it was found that test bars of aluminum bronze show rather constant characteristics and that just a bit of care, required in any foundry work anyway, suffices to produce good castings. So today, a large number of foundries can and do produce decent castings in aluminum-iron bronze; and they do it with exactly the same equipment and men who used to work with other copper-base alloys.

33. Another wrong idea in the field of aluminum bronzes, pertains to the maximum amount of aluminum that can be used without reducing the ductility to zero. That limit was previously indicated as situated near 10.2 per cent aluminum. Today, bronzes are used with 11.5 per cent aluminum and still a ductility of 5 per cent minimum is retained—far more than the best grey iron might possess. Special castings carry even as high as 13.0 per cent aluminum with 4 per cent iron. They possess no ductility in the cold state but are by no means brittle. Cast chisels of this material stand plenty of hard usage under the hammer and do not chip off.

Nickel Silver

34. Among the older alloys that were used in the wrought form for over a century and are now being used in the foundry, we may count nickel silver. Its white, or nearly white, color is unique among copper-base alloys and castings made of it are used mainly in conjunction with wrought nickel silver or heavily nickel plated brass. Essentially, these do not differ much from the wrought alloys in their composition, though they do not vary so much in their nickel content. The latter usually is maintained at 15 per cent and there is nearly always about one per cent tin and 3 per cent lead present. The composition of commercial ingots is so analogous to that of yellow brass that a name like "nickel whitened yellow brass" might be entirely appropriate. In both cases, the presence of tin and the high lead content are quite unwarranted, unless they come in during the process of compounding the charge from various scrap. So much for the old copper-base alloys used in the foundries for the past 20 years at least.

NEW ALLOYS

Silicon Bronze

35. The new alloys form a group of what we might call silicon bronzes, except that the name was in use long ago to denote a low tin bronze wire containing a few hundredths of one per cent of silicon. The cast alloys carry usually about 4 per cent silicon plus one per cent manganese or 2 per cent iron. The presence of one of these last two elements is quite essential, because they (Mn or Fe) introduce profound structural changes in the positive direction and considerably improve the ductility. While not stronger than the best gun metal, they are more reliable and might be cheaper too, were it not for the necessity of using virgin ingots.

Aluminum-Silicon Bronzes

36. To the same group belong aluminum-silicon bronzes with 5 per cent aluminum and 3 per cent silicon. Considerably stronger than the silicon bronzes, they are not half as ductile. Considering the old prejudice against aluminum bronzes, one might expect this alloy to be quite troublesome, but it is not, and the same applies to aluminum bronzes.

Heat Treatable Alloys

37. The youngest members of the foundryman's list of alloys are represented by the heat hardenable compositions. We meet

here the binary copper-beryllium alloys, the ternary alloys of copper with nickel plus either silicon or beryllium or tin, ternary alloys of copper with cobalt plus silicon or beryllium, ternary alloys of copper with chromium plus silicon or beryllium; in short, some eight alloys. Of these, the beryllium-coppers with 2.3 per cent beryllium develop the highest hardness and strength (over 350 Brinell), the copper-nickel-tin alloys come closely enough (near 250 Brinell), then come the copper-nickel-silicon or copper-nickel-beryllium alloys with 190 Brinell, copper-cobalt-silicon or copper-cobalt-beryllium alloys with 150 Brinell and copper-chromium-silicon or copper-chromium-beryllium alloys with 105 Brinell.

38. The last three groups are far superior to any other alloys in the matter of electrical and heat conductivity and while not exceedingly hard at normal temperatures, maintain their hardness at a superior level in comparison with other alloys at temperatures up to 450°C. (840°F.). Of these three groups, those based on copper-chromium are the most conductive and most reliable, due to an almost complete absence of gas porosity and of intergranular shrinkage.

Conductivity Copper

39. Finally, we must mention the non-alloyed, although alloy-treated, high conductivity copper. Not many foundries are producing it because the nature of the process is not clearly understood. The fact that most of the commercial copper ingots, as well as pure wrought copper scrap, contain up to 0.04 per cent oxygen and are gassy, militates against the making of copper castings. Nevertheless, certain foundries are making conductivity castings of 85-90 per cent conductivity as a matter of routine and these castings are quite free from gas-caused porosity. This fact serves to bring their tensile strength to 30,000 lb. per sq. in. and over—close enough to the strength of annealed wrought metal and their ductility is always above 40 per cent, also approaching that of the annealed metal.

40. The essential facts making the manufacture of conductivity castings possible are the following:

1. Copper does not oxidize much on remelting.

2. It absorbs plenty of gas when just molten, but loses the major part of it if brought quickly to 1300°C. (2370°F.).

3. Removed from the furnace and treated with a small amount of a boronic flux or a copper-silicon-calcium deoxidizer, it becomes incapable of keeping the dissolved gases in solution.

4. The amount of the deoxidizer used, must be less than that required to eliminate even as little as 0.02 per cent oxygen, because the addition does not function as a deoxidizer at all.

5. Allowed to stay and cool down to the proper pouring temperature, molten copper loses almost its whole gas content and excellent castings result.

41. In conclusion, we might say that of the six engineering tin bronzes, the four high lead bronzes, five high zinc brasses (including three manganese bronzes), four aluminum bronzes, three silicon bronzes, conductivity copper, nickel silver and eight heat hardenable alloys—a total of 32 alloys—none has become obsolete and all are finding commercial applications.

Light Alloys

42. The situation is quite different in the field of light alloys. The use of aluminum or magnesium-base alloys depends essentially upon their low specific gravity and their capacity to form a protective skin or some complex hydroxide. The necessity for alloying arises from their comparative softness and low elastic limit in the unalloyed state.

Methods of Increasing Hardness

43. The hardness of metals can be increased in two simple ways. The first method is to add elements as will form a solid solution in the main metal. Here, the increment of hardness per unit of element added is quite high, but there is no proportionality between effect and cause. In the second method, elements are added that form a secondary constituent. Here the increment of hardness per unit added is low but a proportionality exists, and a large amount has to be added before the elastic limit is thoroughly raised.

44. There is a third method but it is more complicated, depending upon the phenomenon of intra-granular precipitation and calling for proper alloying plus at least one heat treatment and usually two.

45. Hardening due to the formation of a solid solution has the advantage of maintaining ductility at a high level. When sec-

ondary constituents are formed in the process of crystallization, the ductility invariably drops out of proportion to the hardening developed.

46. Because none of the light metals is capable of forming a wide range of solid solutions, the more desirable method of hardening is rather limited. In the case of aluminum, this is achieved by adding zinc; in the case magnesium, by adding aluminum.

Aluminum Alloys

47. Aluminum zinc alloys enjoyed a superior position in the foundry, at least in Europe, for a long time. They possess a superior hardness, excellent strength, take a high polish, and do not absorb gases in the process of melting. However, the presence of zinc in larger amounts ruins the protective film and makes the castings easily corrodible. As is the case with all solid solutions, both hardness and strength drop rapidly when the temperature is raised. Therefore, the second method of hardening, that of allowing secondary constituents to form, is of much greater importance to the aluminum foundry.

Elements Forming Secondary Constituents

48. A large number of elements can form such secondary constituents. We may list the commonest: (1) copper, (2) manganese, (3) nickel, (4) titanium, (5) chromium, (6) iron, (7) magnesium, all of which form aluminides, and (8) silicon or (9) beryllium which remain elementary; *i.e.* form no compounds. All these elements, when added to aluminum to make castings, cause it rapidly to lose its high ductility and the problem is to use only those elements, and in such quantities, that will produce a considerable amount of hardening before true brittleness develops.

49. This latter usually takes place after an ultimate strength of 22,000 lb. per sq. in., equivalent to a 100 per cent increase in the strength of good commercial aluminum, is reached. No aluminum alloy, used in the shape of an as-cast casting, possesses over 1 per cent ductility when this figure of tensile strength is attained. The combination of mechanical characteristics so obtained is, incidentally, a perfect double for that of the usual gray iron.

50. When a ductility of at least 5 per cent is desired in the as-cast casting, it is necessary to limit the alloying to the point where only 17,000 lb. per sq. in. strength is attained. It always

must be borne in mind that the capacity of forming a protective film must not be inordinately impaired. There can be no talk about keeping it fully intact!

51. There is also another point to be considered. The secondary constituents that provide the hardening effect, differ greatly from aluminum in their specific gravity and their melting point. The chances for segregation are very great and, while spots of high hardness may occur, the casting is likely not to be uniform. Therefore, the amount of the secondary constituents has to be limited so as to avoid undue segregation.

52. These considerations immediately eliminate the use of any large amount of manganese, iron, titanium, chromium, magnesium or nickel and the following ranges of compositions are available if one wishes to get into the strength range of 17,000 to 22,000 lb. per sq. in.:

From 1 to 2 per cent manganese
From 2 to 6 per cent nickel
From 6 to 12 per cent copper
From 3 to 6 per cent magnesium
From 5 to 20 per cent silicon

53. Iron, chromium and titanium tend to segregate badly and should not be present in amounts of more than 0.5 per cent.

54. Indeed narrow is the range of elements and their respective amounts that may usefully be added to aluminum in the foundry. The whole range above mentioned is now covered by known combinations but only a still narrower range is in actual use; namely

Aluminum-copper with 8, 10, and 12 per cent copper—Three alloys

Aluminum-silicon with 5, 10, and 18 per cent silicon—Three alloys

Aluminum-magnesium with 4 per cent magnesium—One alloy

Aluminum-manganese with 2 per cent manganese—One alloy

Eight alloys used in the as-cast state form the whole field of existing compositions and only four of them (8 and 10 per cent copper, 5 and 10 per cent silicon) are in wide use.

Heat Treatable Ternary and Quaternary Aluminum Alloys

55. There has been no dearth of effort to produce ternary and quaternary combinations and in this manner many more alloys, proprietary and otherwise, have become known. Aluminum-copper-silicon alloys were bidding fair for a wide field of uses, but their advantages are offset by the "complications" arising from their use.

56. A wider range of alloys becomes possible if it be agreed to subject the castings to a process of heat treatment. Practically all the binary alloys can be heat treated for increased hardness but an improved ductility is achieved only in those castings that contain very little gas, and, unfortunately, all the binary alloys, except the obsolete aluminum-zinc alloys, tend to absorb gas though a straight 99.5 per cent aluminum does not do so.

57. Better results are obtained if the ternary or quaternary alloys contain some magnesium. A small amount of it assists greatly in developing high strength and hardness plus an improved ductility in aluminum-copper, aluminum-silicon, aluminum-copper-silicon and aluminum-copper-nickel alloys, thus yielding four groups of heat hardenable alloys. In none of the four groups do we possess specific compositions endowed with a set of most favorable characteristics, and a further classification therefore is impossible. The following combinations can be considered as representative:

10 per cent copper and up to 0.20 per cent magnesium
10 per cent silicon and up to 0.3 per cent magnesium
4 per cent copper, 2 per cent nickel and 1.5 per cent magnesium

58. In the last named alloy, the nickel content might be reduced or increased by 1 per cent without much of a change and the magnesium could be reduced to 1 per cent to an advantage. In fact, though this alloy enjoyed quite a boom some time ago, it did not result from a thorough-going investigation of all combinations possible. While varying the copper between 2 and 8 per cent, the nickel between 0.5 and 3 per cent, the magnesium between 0.5 and 2 per cent would have required making not over 126 heats plus perhaps the same number of verifying heats, the renowned "Y" metal, otherwise called magnalite, was not born, but like Topsy just "grewd."

59. We cannot pass without mentioning here the modified aluminum-silicon alloys, containing about 13 per cent silicon and, when treated with sodium and carrying not over 0.3 per cent iron, yielding strong and chemically superior castings with a strength of nearly 30,000 lb. per sq. in. and an elongation of 5 per cent. However, the alloy is likely to be porous and cannot be made from commercial secondary aluminum with its 1 per cent iron plus. For these reasons, and others, it does not seem to form a standard alloy in any of the commercial foundries.

Magnesium Alloys

60. Magnesium is even less capable of forming a great variety of useful casting alloys than aluminum. The binary magnesium-aluminum alloys, for some reason, fall short of requirements and other elements, including copper, manganese and zinc, are added to harden the casting by the method of forming secondary constituents.

61. These complex magnesium base alloys, called "Dow" alloys in this country and "Electrons" in Europe, are too young yet to make possible a proper classification. Most probably one composition forms the bulk of the consumption, others being made when called for, if and when a reasonable assumption is made that the variation will best fit the specific need.

Nickel Alloys

62. Lastly, we shall mention the high nickel alloys, "Monel" and "K-monel," which are fully within the reach of any capable and determined foundryman. The first, a natural alloy with 25-35 per cent copper plus variable amounts of iron, manganese, silicon and a bit of carbon, is sufficiently well known to require no additional description. The "K-monel" is obtained by adding to the monel up to 5 per cent aluminum and subjecting it to a solubilizing high temperature (900-1000°C. or 1650-1830°F.) first, and a hardening treatment (500-600°C. or 930-1110°F.) later. It develops a hardness of over 300 Brinell and is highly resistant to corrosion.

63. No other high nickel alloy of today is within the means of the regular non-ferrous foundry.

SUMMARY

64. Summing up our list we find the following:

(1) Six engineering tin bronzes (red brass, admiralty, steam metal, gear metal, 14 per cent tin bronze, and bell metal).

(2) Four bearing-lead bronzes (10-10, 10-20, 5-30, and 60-40).

(3) Four aluminum bronzes (9 Al-3 Fe, 10.5 Al-4 Fe, 11.5 Al-4 Fe, and 13.0 Al-5 Fe).

(4) Three silicon bronzes (4 Si-1 Mn, 4 Si-2 Fe, and 5 Al-3 Si).

(5) Five brasses (Muntz metal, three manganese bronzes and yellow brass).

(6) Eight heat hardenable copper-base alloys.

(7) Straight high conductivity copper.

(8) Nickel-silver.

(9) One bronze of indefinite or obscure composition, as used for statuary work.

(10) Four aluminum-base alloys used in the as-cast state.

(11) Three aluminum-base alloys used after heat treatment.

(12) One aluminum-silicon alloy used after a modifying treatment.

(13) One magnesium base alloy.

(14) Two nickel base alloys

making a total of 44 alloys.

Part II — Metallurgical Science and the Foundry

65. Having taken an inventory of the individual alloys which the foundryman can use now, we shall go further and see to what extent science has helped, is helping, and may help the foundryman, and to what extent he might engage the services of science in the future.

THE EQUILIBRIUM DIAGRAM

66. Investigation in the natures of alloys was not the first step in the approach of science to practical metallurgical problems. This was antedated by the advent of applied mechanics with its study of mechanical characteristics. This latter step started about 1825 in its application to cast iron and provided the first indications that metals and alloys might and should be improved. The result was that progressive foundrymen here and there began to look empirically for such improvements, even though not a single leading thread was in their hands.

67. The way became clearer when, one after another, came the studies of equilibrium diagrams for the carbon-iron series, then for those of copper-tin, copper-zinc, and copper-silver series; all of which provided combinations of all the essential structural features to be had in the general field of alloys.

68. Those first studies provided a good explanation of mechanical characteristics on the basis of structural relationships. They showed that a metal becomes specifically hardened in the measure it takes other metals into solution in its crystals and that the appearance of a secondary constituent—especially between the grains—brings about an *additive* increase in hardness and a rapid loss of ductility.

69. The first studies of alloy constitutions were wide in range; *i.e.* embracing the whole series in the case of binary alloys, but by no means sufficiently precise. It took nearly ten independent investigations over a period of 40 years, clearly to establish the constitutional features of copper-tin alloys, and nevertheless, there remain certain features in the most important field of the alpha solid solutions which the diagram does not explain and which call for a variety of hypotheses.

70. At any rate, we possess today rather complete information on the constitutions of the following binary alloys:

(a) Copper with (1) zinc, (2) tin, (3) nickel, (4) iron, (5) aluminum, (6) silicon, (7) antimony, (8) arsenic, (9) phosphorus, (10) lead, (11) silver, (12) chromium, (13) magnesium, (14) cadmium, (15) manganese, (16) beryllium.

(b) Aluminum with (1) copper, (2) magnesium, (3) manganese, (4) silicon, (5) nickel, (6) iron, (7) zinc, (8) chromium, (9) cobalt, (10) silver, (11) lithium, (12) calcium, (13) antimony.

(c) Magnesium with (1) aluminum, (2) copper, (3) manganese.

71. Ternary and quaternary alloys form a vastly more difficult field for constitutional investigations. Not even one hundred experimental melts, subjected to thermic analysis, can yield enough data to permit a satisfactory sketch of constitutional situations and changes for a complete ternary system. And as far as the industrial alloys are concerned, the variations of composition from one experimental heat to another must be quite small or a precise picture will not be obtained. Should we desire, for instance, to get a good constitutional diagram for copper-zinc-aluminum alloys in the region embracing 15 per cent aluminum and 55 per cent zinc, a problem would arise of investigating some 120 basic and probably 30 more verifying heats. It can be done of course, but so far no such decisive effort has been made. That is why our knowledge of constitutional diagrams of ternary and more complicated alloys in the domain of the copper-base systems is so sketchy as to be practically useless.

72. The situation is better in the case of aluminum-base alloys. Not only are fewer experiments necessary, but in addition, the temperatures of solidification and transformation are so low that the technique of the thermic analysis is greatly simplified. Due to the efforts of the aluminum producers in this country and the British investigators, we now possess reliable diagrams for the aluminum-copper-silicon, aluminum-copper-nickel, aluminum-copper-magnesium, aluminum-iron-silicon, aluminum-magnesium-silicon, aluminum-zinc-magnesium through ranges extending far beyond the possible industrial alloy compositions.

73. All these results of well investigated series and systems are of great value to the makers of the wrought alloys. The latter

have to pass through so much kneading and heating that, as far as their constitutional characteristics are concerned, they are in a complete state of equilibrium, or else in precisely that distance from it as may suit our needs.

74. The situation is different in the case of the foundryman. Most of his products are never heat treated and never reach the equilibrium state because the solidification and cooling proceed at too high a speed. In addition, the distance from the equilibrium state depends on the size of the casting and its thickness.

75. Further still, hardly an alloy used in the foundry represents a simple binary or ternary system. We must introduce lead wherever it does no harm for the purpose of enhancing machinability, and as much of the cheaper elements as the job will stand, to cut down the costs of production. Who will make bold to say that a commercial bronze with 8 per cent tin, 4 per cent zinc, 3 per cent lead, plus a variable though small amount of iron, antimony, nickel, arsenic and last but not least a substantial amount of occluded gases, will not show a vast constitutional difference from that of a pure mixture of 88 per cent copper, 8 per cent tin and 4 per cent zinc made in an electrical resistance furnace?

76. It must be stated emphatically that, in their present state, the equilibrium diagrams mean very little to the practical foundryman no matter how progressive and scientific he may be. Of course, an educated foundryman must have before his eyes a fairly accurate picture of every diagram of importance in his field; he can not expect, however, to make a better use of them than appraising and discarding any wild scheme of "alloy improvement."

THE MICROSCOPE

77. The present day metallurgical microscope owes its existence in an equal measure to the Frenchman, Le Chatelier, and the German, Martens. The first conceived the instrument as an auxiliary apparatus to be used in studying constitutional diagrams via structural observations. The second intended it mainly for the study of the causes of changes and imperfections in steels. Today, however, the microscope occupies the first place in the examination of constitutional features where its precision exceeds that of the thermic analysis. Of course, its main use covers the field of such structural characteristics as grain size, inclusions, dendritism,

anomalies, grain distortion, etc., which can not be read at all from the best constitutional diagram available.

78. Here again the maker of wrought alloys is decidedly at an advantage. A few specimens taken from a batch of steel rails will tell the metallographer whether the metal abounds in inclusions or not, or whether a batch of sheet brass, intended for a deep drawing job, has the proper grain size.

79. The non-ferrous foundryman who wishes to use the microscope has a thankless problem before him. Even the starting point; *i.e.* getting a well polished surface, is fraught with difficulties. This is because even the best casting will possess a certain amount of gas holes that become collecting pockets for the polishing medium and the particles of the latter, released on the reversal of the direction of polishing tend to produce new scratches.

80. While examining the polished surface in the unetched state, he is likely to notice a large number of dark spots. What do they represent? Are they holes or inclusions? If holes, are they gas holes or shrinkage cavities? If they are inclusions, are they grainlets of oxide, sulphide or nearly eroled grainlets of lead? To decide on these questions requires almost a prophetic gift, because both holes and inclusions are equally dark, and because most of the inclusions will be pulled out in polishing unless the polisher possesses the patience of a saint, and a true love for his humble work.

81. Furthermore, the distribution of holes and inclusions (the latter, by the way, are devoid of any importance unless extremely large) varies from spot to spot and no fair average can be obtained. So, while the need for a microstructural examination arises mostly in the case of failures or rejections, it is quite possible for the customer to locate a spot that is all holes or all dirt, and for the maker to find very close to it another spot of a fairly uniform and sound structure.

82. The foundryman can gain even less valid information from the examination of etched surfaces. In the case of tin or silicon bronzes and some aluminum bronzes, the matrix will be formed of highly cored grains among which he will see the secondary constituent—a eutectoid in all three cases. Of the grain size, the microscope will tell him nothing because the smallest grain will be more than 1 mm. across and consequently will occupy the

whole field of vision at any magnification. The grain-size in castings has to be ascertained from a macrograph if it is deemed to be of importance.

83. Now, what can a foundryman say when he sees a dendritic or cored structure of the individual grains? He knows beforehand that the grain must be dendritic because the equilibrium state is not reached and the central part of the grain differs chemically from its outside. He knows also that the intensity of the coring effect depends upon the etching reagent and the conditions of etching, which cannot be so far precisely controlled. Therefore, he can not even pass judgment as to the possible inhomogeneity grade of the individual grain—not that it would help him much were he able to do so!

84. What can he say when he observed the delta eutectoid in admiralty gun metal, for instance? Its amount will depend on the conditions of cooling and, to an extent, upon the lead content. It will be larger in the test piece than in a thick part of a casting. Logically, the increased amount ought to mean more hardness, but less strength and ductility. But the tool of simple logic fails here. A good standard test bar may possess 15 per cent of the eutectoid by volume and show 45,000 lb. per sq. in. strength plus 35 per cent elongation, while a bar taken from the big casting, simultaneously produced, may carry only 3 per cent of the delta constituent, but show only 37,000 lb. per sq. in. strength and 20 per cent elongation.

85. Of course, the microscope may produce visual evidence in some extreme cases. It may show that the casting was bad when any spot examined exhibits a network of large and small holes. It may show that an error in compounding brought in too much tin if the amount of the eutectoid observed reaches 30 per cent instead of the normal 5 per cent. Other methods, however, such as density determination and chemical analysis, will explain such cases in a more definite manner, inasmuch as they can yield true averages which micro-investigation cannot do.

86. Suppose that a metallurgist wishes to apply the microscope to the study of manganese and aluminum bronzes, both of which consist essentially of a mixture of the alpha and beta phases plus some bluish looking constituent of an oval form due to the presence of iron.

87. Having cast a propeller of manganese bronze, the foun-

dryman chisels off the cast-on test bar and uses a piece of it for the microstructural investigation. If he has previously had considerable experience in examining such samples, he will be able to tell right away whether the ratio of the two constituents is a normal one, and he may even get a nice picture of the "blue constituent" in addition.

88. But the foundryman and his customer are not interested simply in looking at "nice" pictures. If they ask the metallographer for an interpretation, at the present state of our knowledge he cannot possibly undertake to say how the distribution of the grainlets of this blue constituent may affect the characteristics of the casting.

89. On the other hand, his estimate of the alpha/beta ratio is of very little consequence. A look with the unaided but experienced eye upon the fracture, a plain Brinell test or a density determination will indicate the variation of the ratio in question from the normal with equal if not higher precision, and each of these tests is cheaper and quicker.

90. In the case of aluminum bronze, the metallographer might like to know whether the casting has become defective due to "self-anneal." This self-anneal corresponds approximately to the phenomenon of the dissociation of the beta phase. The initial stage of this dissociation takes the shape of an acicular structure of the beta grain.

91. However, the chemical differences that bring out the acicular structure in the individual beta grain are not half as strong as those reigning at the boundaries of the alpha and beta phases. Therefore, only an exceptional etch might show the deterioration of the beta phase. On the other hand, a hardness test and a bend test will discover the presence of such deterioration immediately. However, such deterioration can never take place in a casting which does not carry over 9 per cent aluminum and contain some iron.

92. The situation is still more negative in the case of heat hardenable alloys. The hardening precipitation is a submicroscopical phenomenon and no lens will tell the difference between the homogenized and the properly heat-aged metal. The microscope may show, however, whether much of the hardening compound has

remained undissolved and coalesced along the grain-boundaries, so causing a loss in ductility. Unfortunately, the remaining grainlets are always very small and their presence can be ascertained at higher magnifications only. Only a true artist in metallography likes to work with 4 mm. or the 2 mm. oil immersion lens.

93. In short, the production foundryman hardly can expect to make good use of his micro-equipment. To him, it is merely an expensive toy which comes in handy for preliminary development work on the introduction of a new alloy, the properties of which are not clear. But such cases are rare.

THE X-RAY APPARATUS

94. The possibility of recognizing very unsound spots in pieces of metal photographed in the light of the transmitted x-rays has long been recognized. The holes drilled in metallic keys produced strong shadows in the very first photographs taken by Professor Roentgen. Nevertheless, the method remained only a possibility as long as the old soft tubes were used in connection with low voltage excitation.

95. Higher voltages came into use after the development of methods for studying the structures of crystals of metals and alloys. From there, another step produced tubes of very high voltage excitation, the hard rays of which are capable of penetrating heavy layers of metals and telling the story of their local soundness.

96. Crystal analysis still remains a purely theoretical issue for the foundryman. The very fact that the grains are mostly denditric and that the x-rays produce a picture of the raw averages of atomic distances make the application of this branch of study to foundry problems quite impossible.

97. The situation is only slightly more favorable in the case of radiography. The method may be used advantageously where a large number of small simple cast parts is produced in a batch. Symmetrically disposed on the table of the apparatus and photographed in the transmitted x-light, the unsound pieces will differ sharply from the sound ones.

98. Even simple, but large, castings cannot be efficiently radiographed in one shot. The farther from a perpendicular attack by the rays a part of the casting is located, the less intensive the total attack per unit of area (in the second power of the distance

from the anti-cathode), and the thicker appears that part in the direction of the transmitted x-rays. Therefore, an evenly illuminated picture of a long cast bar or of a large cast plate is difficult to secure. Only in those cases where it is suspected that a given spot is unsound, and *irregularly unsound* in addition, will the x-rays tell the true tale. On the other hand, an evenly porous casting of gun metal of the specific density of 8.65 (against the maximum of 8.85) will not produce a much different picture from an evenly porous casting with a density of 8.50. Yet their physical characteristics will be vastly different.

99. Bronze gears or gear blanks can be investigated with the x-rays to a great advantage, for here the method permits of the disclosure of any shrinkage cavity of importance. They might also help to ascertain whether a given design of a large wheel does not lead to the formation of shrinkage cavities at the individual joints of the spokes and the rim. But those are only secondary troubles in a non-ferrous foundryman's life; general porosity and intergranular shrinkage are his real enemies and here x-rays are of no avail. To think of investigating by x-rays a large casting for a propeller shaft's jacket weighing about 6000 lb., or the propeller itself is hopeless. So we are forced to conclude that even the x-rays, one of the most powerful weapons of metallurgical science, can be of but little help to the practical foundryman.

CHEMICAL ANALYSIS

100. Except for a new Rockwell hardness tester or an old Brinell press, the chemical laboratory is probably the most frequent adjunct of the non-ferrous foundry. Its functions there are manifold but can be subdivided into two classes of services: the commercial and the technical.

101. While most of the foundries have to use some virgin metal to make up a charge of a given composition, commercially the best, and actually the more rational way, is to buy ingots of the proper type from a reliable scrap refiner. Most of the latter are quite capable of producing ingots within the range of the chemical specifications and very frequently the refiner makes sure before making a shipment of such ingots that a properly handled heat will yield test bars exceeding the minimum requirements of the various standards.

102. But not all refiners are on the same level and not everyone's invoice can be trusted. Therefore, the foundry laboratory has to analyze now and then the incoming batches of ingots to make sure that a shipment of gun metal really averages 10 per cent tin plus or minus a few tenths of one per cent. Such services we might call commercial, as the technical side of the job does not form the first consideration.

103. Where alloys are prepared and ingoted right at the foundry, the laboratory may be called upon from time to time to check the correspondence of the ingots made to the charge ordered. This, too, may be considered in the nature of a commercial service. It provides a control of the conscientiousness of the man making up the charge in his relations to his employers. Even here, a chemical analysis is not usually called for, unless a preliminary examination of the filed down surface and a hardness test indicate a suspicious deviation from the normal.

104. Far more important might be the services of the chemical laboratory in helping to clear up cases of failures in castings. The personnel of the chemical laboratory is probably well enough trained to determine precisely the amounts of tin, lead, copper, aluminum, and iron present. In fact, unusual precision in the determination of these elements is not necessary. It is of but little importance whether a leaded admiralty gun metal contains 10.2 or 9.7 per cent tin, 2 or 1 per cent zinc, 0.5 or 0.8 per cent lead, 0.2 or 0.3 per cent iron (from the stirring bar). So, should the chemist be careless enough to lose 5 cc. of his second washing water, or should he drop a tiny bit of the precipitate of metastannic acid, the error will be too small to vitiate the analytical results.

105. On the other hand, the failure of an aluminum bronze might be due to the presence of small amounts of arsenic in the copper scrap used. A manganese bronze may suffer badly if more than a trace of antimony gets in the metal and a leaded bronze may be ruined by the presence of 0.05 per cent aluminum (as the accepted opinion says). But will the foundry's laboratory be able to give a quick and precise answer to such questions? We venture the opinion that a laboratory so equipped and manned would prove too costly for routine work.

106. Perhaps a far more important service could be given by the laboratory if it were able to tell promptly and precisely how

much gas a casting contained and what was the nature of such gas. But is there a foundry anywhere today which is able to cope with problems of this type, otherwise than tentatively?

107. The steel foundry handling tons of metal at a time can now and then establish a rapid control (colorimetric) of the essential elements present. For commercial and technical reasons, the non-ferrous foundry cannot allow the few hundred pounds of molten metal in the crucible to linger in the furnace for half an hour extra even if it were possible to return reliable data on the composition of the metal in that short time. With the pouring done, there is no need to analyze the metal if the castings are good while a post factum analysis in cases of failure does not help much. A controlled remelt of a few pounds of the scrap, followed by simple mechanical tests that do not take one half hour, will tell definitely whether it was the composition that was at fault. After that, the finding of the true troublemaker will be the task of the umpire. And at any rate, castings fail mostly on account of porosity and misplaced shrinkage, and cases where the composition is at fault are rare.

MECHANICAL TESTING

108. Tests of a mechanical nature can be subdivided into four classes:

(1) Those requiring no machinery and yielding crude qualitative results,

(2) Simple hardness tests using the Brinell or the Rockwell apparatus,

(3) Specialized tests for fatigue and impact,

(4) The usual static tests.

The highly specialized creep and abrasion tests are still in their infancy anyway and foundrymen will not be interested in them for years to come.

Observation Results

109. The crude qualitative tests form the best method for the study of cast metals. A small sand cast sample of not too plain a shape if thoroughly belabored with the hammer, the chisel and the file will reveal imperfections that no other test will show. Scores of fractures can be examined in a few minutes. And if the foundry

manager and the test engineer is willing, the breaking of one actual casting selected at random will tell a great deal about the general and local porosity, shrinkage cavities and shrinkage cracks, presence of dirt, etc.

110. The only trouble with such tests is that they are likely to become perfunctory. Their results cannot be expressed in a report of 300 plain words and the momentary impression tends to evaporate so that succeeding tests do not benefit from the results of the past ones.

111. Here we come to the conclusion that a well trained memory and a strong determination to use it are the most essential and unalienable tools of a progressive foundryman. He, or the man who works for him, must train himself to memorize and classify in his mind the results of these crude but rapid and most revealing tests.

Hardness Tests

112. The second class—the hardness tests—are much beloved because they are so cheap in execution and yield figures, not mere impressions upon the unwilling brain. In certain ways, the hardness test is really worth while. The surface of a casting of a given composition will always yield closely coinciding hardness data. If, in one case, the hardness drops from 70 Brinell normal to 65 or rises to 75 the composition undoubtedly differs a lot from the desired one. Or, if a heat-treated casting of copper-chromium shows a surface hardness of 95 instead of 105, a goodly loss in the recovery of the chromium added becomes apparent.

113. However, surface hardness is no guarantee that the hardness of the deeper layers will be of the same order or will even vary in a uniform manner. In many an alloy of the tin bronze type, the hardness 10 mm. below the surface may drop by 50 per cent. Porosity and shrinkage inside the casting produce but remote effects upon the hardness of the surface. The latter, to the depth of one or 2 mm., acts as if it were a shell having nothing in common with the metal beneath.

114. Another dubious point presents itself when different types of hardness tests are being used. The finer ones, like the Vickers pyramid test, are hardly applicable to the study of cast metals, although they are nearly ideal when applied to wrought

articles. But even the two rougher ones, the Brinell test with its 10 mm. ball and a 500 kg. load, and the Rockwell "E" with its ⅛-in. indentor and a 100 kg. load are by no means comparable, when the inside of the casting is studied. The impressions cannot be made at the same spot—that much is obvious. Even taking ten Brinell readings and ten Rockwell readings, averaging each and correlating the results produces no precise information. The averages of the next groups of ten readings may be vastly different from the first ones and the differences do not have to run in the same direction.

Fatigue and Impact Tests

115. It may seem strange to take up fatigue and impact tests before discussing the static tests, but this is done here so as to dispose early of another unpromising subject. Fatigue testing belongs to those experimental methods that rapidly capture human fancy. The very fact that one has to run scores of tests at various loads and for periods covering months strikes man with awe. There was a time when fatigue tests were believed to lead us to the true understanding of the nature of permanent deformations and incipient failures.

116. Foundrymen were never thoroughly interested in fatigue tests and rightly so. The test is in itself too conventional, too dependent upon a given shape of the test piece and upon the apparatus used to permit of far-reaching conclusions. The raw bars purposely cast to be machined into fatigue specimens under ideal conditions, might produce important data as to the stress at which the first slip starts for a given alloy composition. A gun metal charge made of virgin materials of high purity in an atmosphere of argon or helium and cast with the maximum precaution into a series mold of dry sand so as to yield 30 to 40 round bars for further machining, might provide us with very valuable data on the basis of numerous carefully checked fatigue tests.

117. But such ideal gun metal may have but little in common with the alloy used in the most progressive foundry. Nine times out of ten, a tiny pinhole will occur exactly where the maximum strains are supposed to take place and the results of the fatigue tests will show no regularity whatsoever.

118. Almost the same thing applies to the impact test. It is not too difficult nor too costly to cast a number of square samples

about 12 mm. across and 100 mm. long, to machine them into regulation test bars provided with regulation notches, while breaking them in the impact machine takes practically no time at all.

119. The results of such tests, however, will vary greatly unless the original samples were machined and notched in a precisely identical way. In addition, these results may not correspond in any definite degree to the impact resistant qualities of the casting itself. So while it is fairly reasonable to expect that the actual casting, while inferior to the separately cast tensile test bars, will possess an average tensile strength equal to 80 per cent or more of the strength of the test bar, actual experiments have shown that impact strengths may vary within the same casting by 70 per cent, according to the location of the notch.

120. All told, we are compelled to conclude that the specialized tests for fatigue and impact resistance, like others in the same class, do not interest the foundryman. Too many are the conditions to be met if the results are to have a scientific value and even the best foundry practice imaginable cannot derive elucidating conceptions from such results.

Tensile Tests

121. There remains now the old, time-sanctioned and logical test of the standard bar in tension. It is not an absolutely faultless test and there still remains the conventionality of the dimensions of the test bar, the arbitrarity of the elongation figure. But even so, it is far superior to others, for it has a definite physical significance and, with due allowances for other factors, it defines well enough the given batch of metal, if not the casting made of it.

122. Only the small foundries can afford to do without a tensile testing machine for it is always preferable to conduct the acceptance test right in the foundry or to forestall its results by running one's own, even if not an officially recognized test. However, in most foundries, the tensile test is resorted to only when contracts make it obligatory, or when a member of the staff waxes ambitious and tries to develop a new alloy or a new procedure in handling an old one.

123. It would be obviously impossible even for the best foundry to cast test bars of each heat and subject such test bars to machining and testing. It requires at least two test bars to arrive

at a substantiated conclusion and not even a heat of 600 lb. would stand the expense involved. So there can be no question of running tensile tests in a regular way, day in and day out. There is, however, one point where the foundries fail to derive the maximum benefit from the tests actually performed. That is in the extreme formality with which such tests are run. So, because no standards demand or indicate the range in which the figures of area reduction, elastic limit and elasticity modulus of a given alloy must fall, these data are usually not considered. And once the test is performed and the piece broken, the chance of getting these data is gone forever.

The Value of Records

124. There is one more dark point of which we shall talk. The written records of the tests are of course not destroyed right away but go into storage for a time, *exclusively* for the purpose of keeping a recorded evidence handy in case arguments arise. The old records accumulate dust and are never brought to light for curiosity's sake.

125. It requires no proof to state that any sizable foundry, catering to exacting customers, has in its archives thousands of records of tests run on alloys of important nature: admiralty gun metal, steam metal, aluminum bronze, manganese bronze, etc. A simple statistical examination of the various characteristics would undoubtedly produce very instructive frequency curves showing the degree of probability that a certain feature will fall within a given range. A more complicated examination, in which the results would be correlated to the source of the ingots, the day of the week, the foundry boss in charge, the nature of coke, etc., might produce valuable indications and so eliminate some intrinsic causes of failure, or enhance those of success.

126. Wherever conditions permit, a foundry should not be satisfied to buy some machine that a steel rolling mill or a college is ready to junk. The tensile testing machine is frequently the final arbiter and should be equipped in the best manner possible. A precise extensometer, or better still, a good recording device is essential. Its indications may not interest the customer or the judge, but it costs little to make them and the larger record calls for no more food per day than the abbreviated one.

Summary of Tests

127. Summing up the preceding matters we wish to state again that with the exception of the crude qualitative tests and the tensile testing—if properly conducted, recorded and examined —no other scientific tool nor theory (constitutional diagrams, crystal germination, etc.) can be of great assistance to the foundryman in his task of turning out high quality products at a conceivable minimum expense. Metallurgical science and the foundry industry may pretend to foster one another, but as a matter of fact, their contacts are rare, mostly unsatisfactory to both and what is worse, cannot be made more continuously satisfactory. It simply is not yet in the nature of things.

128. This does not mean that the foundryman should oppose his metallurgist when the latter asks for various scientific equipment. On the contrary, whenever a company running a foundry is not close to bankruptcy, it should bring in better laboratory equipment, more books and magazines and engage more men with a scientific viewpoint.

129. If this expense serves to make the foundry's staff better educated, more capable of understanding the experimental work of other people and better able to reproduce the experiments, the profits will be invariably there. They may not be directly discernible in dollars and cents and that foundry president makes a big mistake when he expects that installing some scientific equipment will increase his profits within a "reasonable" period of time. It can't be done.

CAN SCIENCE HELP?

130. "But," the foundryman may ask, "my business demands that I continuously improve my products, and that I eliminate failures and rejections to a great extent. Can science or the scientific method help me in that permanent task?" To this question the author answers:

131. Your business depends on two major things—first, a thorough understanding of general foundry processes; secondly, the art of proper management and of individual adaptation. The products of your business cover a wide field of shapes, sizes, and applications. The proper handling of each of these constitutes an art and developing this art to perfection is the task of your individual foundry.

132. But the general processes are a different thing. No individual foundry is rich enough to investigate them all of its own accord; and even if it were, it could ill afford to do so because the costly results would benefit its competitors at no cost to them. Therefore, the investigation and re-investigation of the generalities must be the business of an organization whose results are free to anyone, an organization supported on the basis of proper distribution of costs among all the parties interested.

133. To illustrate that individual effort is unlikely to shed real light on foundry problems, I shall cite the following cases. An investigator decides to find what temperature is best for the pouring of a given alloy. He makes up 12 standard test bar molds and bringing a sizable amount of the metal to a high temperature pours three molds at the highest reading of the pyrometer, three more about 50 degrees lower, three more 100 degrees lower, and the final three 150 degrees lower. Then he proceeds to test the bars, computes the arithmetical averages, draws a diagram, and pronto! the proper temperature point for the pouring of the alloy in question is determined.

134. Another wants to find whether arsenic is deleterious to a given alloy. He proceeds in nearly the same way. He makes four or five charges with an increasing amount of arsenic in each. He pours the molds at the identical temperatures, pulls the test bars, averages the results, draws a diagram, and just as pronto he comes to the conclusion that up to a certain point arsenic is harmless, etc., etc.

135. Nothing is more erroneous than this procedure. It calls for closing our eyes and minds to the various "other" factors that were present but not accounted for. To find the true relationships one has to make hundreds of heats and cast a large number of test bars from each plus a large number of more intricate castings. Only then, could the investigator claim that *in all probability* the incidental influence of those "other" factors was balanced out and that he had obtained a true picture of the influence of the pouring temperature, of a given impurity, or of an intended addition.

136. In other words, a mass production of research data is the only method by which the general processes can be made clear and freed of antiquated or ill-substantiated notions. This is what

science can do for the foundrymen and this is also where the foundrymen can advance metallurgical science.

137. Here is a list of questions which such a research program ought to answer—clearly and completely:

(1) There must be one inexpensive copper base alloy for general use in conditions where neither high pressure nor a continuous friction is encountered. Shall it be a bronze (including the three fives) or a brass? What is the most logical composition of it?

(2) There must be one or two different alloys that are likely to give the highest satisfaction when used for high pressure castings under general conditions (damp air, steam, water, saline solutions). These alloys must also be indicated clearly.

(3) There must be a leaded bronze most likely to stand up as a non-babbitted bearing or bushing material. This composition must be determined and precise information obtained concerning the adjustments of the lead content to the more specific needs.

138. The same procedure should be applied to other alloys as well. There is no good reason why one company should be using an aluminum bronze with 9 per cent aluminum plus 3 per cent iron and another 8 per cent aluminum plus the same amount of iron for identical jobs, or why one company should produce ship propellers of an alloy of 75,000 lb. per sq. in. strength while another goes to 88,000 lb. per sq. in. For one or the other alloy must be better, while the price difference is negligible.

139. This does not mean, of course, that a law should inhibit the customer from asking or the foundryman from offering any alloy which they may desire. But, our knowledge of good and not so good alloys must be standardized on the basis of a profound investigation.

140. Further on, the problems of gas absorption and of intergranular shrinkage have to be tackled in full. We must get a perfect knowledge regarding the nature of the absorbed gases and their amount. We must know how the latter changes in time and temperature and how the potent "deoxidizers" actually do their work. Guess-work and false notions that cover this field must go.

141. We do not pretend to offer here a complete program. It is our desire only to state emphatically that a vast program of mass data research must be inaugurated if the foundry industry is to progress rapidly and systematically, and not by the old processes of groping in the dark and "muddling through" which were used 4000 years ago.

142. Furthermore, we shall add that there is no need of creating a permanent organization for such research. A permanent organization soon becomes lazy because it can afford to defer till tomorrow what it should do today. And it begins to consider itself the owner of the enterprize, having the right to see to it that the available cake shall last longer for the chosen few.

143. This pitfall should be avoided too. The fundamental data upon which the foundries of tomorrow shall proceed with their individual work must be gathered and analyzed by a virile short term organization. A period of two years energetically invested should completely exhaust the problems existing. *The ideal foundry is one that is run with so thorough a knowledge of the fundamentals, and so efficiently, that no testing of any kind becomes necessary.*

DISCUSSION

Presiding: H. M. ST. JOHN, Crane Company, Chicago.

G. H. CLAMER—Listening to this paper I was particularly interested in the last sentence, "The ideal foundry is one that is run with so thorough a knowledge of the fundamentals, and so efficiently, that no testing of any kind becomes necessary." It seems to me if I were to criticize or discuss that particular sentence I would include in the same that the ideal foundry is one that is operated by the ideal human and with ideal raw materials.

In Mr. O'Connor's paper of yesterday I was particularly impressed in noting the many factors having an influence upon the production of castings and the great care that is necessary in watching each step involved. The thought came to me as I listened to that splendid paper that it would be a worthy undertaking for the AFA to compile a list of all the factors involved in the making of castings. I am not unmindful of the splendid work of the AFA in preparing its publication on foundry defects and I am merely proposing amplification thereof. Probably if we were to check all the factors mentioned by Mr. O'Connor in his paper, including the human factor, we might have a list of at least fifty.

[1] President, Ajax Metal Company, Philadelphia, Pa.

MR. CORSON: Of course, I have to agree with Dr. Clamer when he says that my statement concerning the ideal foundry may be exaggerated. All ideals are exaggerations that serve as conning towers along the road of progress. On the other hand, the ideal foundry is probably the most easily attainable goal among other technical ideals. Buying ingots from a reliable manufacturer, practically eliminates the need of worrying about the performance of the metal as such.

The use of properly built furnaces manned by intelligent furnacemen not overburdened by too many units,—and I have seen cases where a practically illiterate young man was handling three revolving reverberatories and five tilting crucibles at the same time, turning out terrible metal but very cheaply—is the second step.

Finally, sand properly cared for and molding done under good supervision form the remaining desiderata that must go into the making of a foundry which might be ninety-five per cent ideal. In fact, it seems that Mr. O'Connor, who read his paper yesterday, came to the point where his foundry is closer to the ideal than the above figure implies. And, if Mr. O'Connor's methods were properly recorded, studied and his molding logics adhered to by other foundrymen, another step toward the ideal might be on hand.

L. W. KEMPF[2]: I think Mr. Corson has done us a real service in attempting to reduce some of the mass of specific information on the making of castings to generalities. The sooner we can come to generalities the better able we will be to make intelligent selections and prophecies. Sometimes, however, the exceptions to the generalities are of paramount importance to the man who is planning to make money in the foundry business.

During Mr. Corson's presentation, I took occasion to look over the section of his paper on light alloys, since I am primarily interested in these materials. His generalities appear to be, in the main, correct. However, I thought it might be worth while to point out a few exceptions because, as I said before, these exceptions frequently mean the difference between profit and loss.

On page 62 the author states, "Iron, chromium and titanium tend to segregate badly and should not be present in amounts of more than 0.5 per cent." Most anyone who has been making aluminum castings for the trade at the lowest possible price to the user knows he can't live up to this requirement. Iron generally in his castings will run higher than 0.5 per cent, not primarily because of the source of his metal but because he wants it higher for the improvement in casting characteristics.

As a matter of fact, something over twenty years ago, when one of the most important problems before this country was the production of aluminum alloy castings for Liberty engines, a great number of cast-

[2] Aluminum Research Laboratories, Aluminum Company of America, Cleveland, O.

ings were being scrapped in the foundry because of cracks. Also, a great number of them were failing after relatively short service in the engines.

A very intensive and highly scientific bit of research definitely established the fact that iron content considerably in excess of 0.5 per cent gave not only a higher degree of toughness in the finished casting but also gave freedom from casting cracks and I doubt if any producer of unheat-treated castings today, particularly from secondary alloys, runs his iron as low as 0.5 per cent. If it happened to get this low, he would probably bring the iron up to 0.75 or even 1.25 per cent.

It is true chromium and titanium do tend to segregate, titanium much more than chromium. There is no occasion for running these elements anywhere as near as high as 0.5 per cent.

Mr. Corson finally narrows the field down to 4 alloys, 8 and 10 per cent copper and 5 and 10 per cent silicon. Of course, these could be narrowed down even further to 8 per cent copper, roughly speaking, and 5 per cent silicon. Ten per cent silicon is practically unused commercially. The 10 per cent copper is used very little for castings from secondary metal. It is used principally for permanent mold piston castings.

On the other hand, this doesn't give us an exact picture of the true set of conditions. There are a good many variations of these two fundamental alloys and to dismiss the entire matter with the statement that the 8 per cent copper and the 5 per cent silicon are the only two common casting alloys is a little bit too wide a generalization.

In the next paragraph Mr. Corson states, "Aluminum-copper-silicon alloys were bidding fair for a wide field of uses, but their advantages are offset by the 'complications' arising from their use." I am not certain what Mr. Corson means by "complications." By far the greatest production in aluminum castings at the present is in A.F.A.* Aluminum Alloy B-13 which ordinarily contains about 7 per cent copper, 1 per cent iron, and 1.5 per cent silicon. The silicon is an important part of the alloy, as anyone can attest who has attempted to produce leak-tight castings from straight 8 per cent copper alloy.

Discussing the heat-treated alloys, Mr. Corson states that better results are obtained if they contain some magnesium. Magnesium is an important element in the heat-treatable aluminum-silicon alloys, but its presence is always accompanied by a decrease in ductility as compared with the magnesium-free alloy. It increases strength only. In the commercial heat-treated aluminum-copper alloy containing 4 per cent copper, the best combinations of strength and ductility are obtained in the absence of magnesium.

Under magnesium alloys Mr. Corson states that "the binary mag-

* A.F.A. Cast Metals Handbook, 1935.

nesium-aluminum alloys, for some reason, fall short of requirements and other elements, including copper, manganese and zinc, are added to harden the casting by the methods of forming secondary constituents." The binary or substantially the binary magnesium-aluminum alloys were used for quite a number of years. I say substantially binary because they contained in addition to 7-8 per cent aluminum only a few tenths per cent manganese. Manganese is added to all magnesium alloys to improve corrosion resistance. Copper is practically never used as a constituent of an alloy with magnesium. It so decreases the corrosion resistance that it is of no practical use. I might state here that practically 80 to 90 per cent of all magnesium castings produced in this country are heat treated. The most common alloy contains about 6 per cent of aluminum, about 3 per cent zinc, and 0.2 per cent manganese.

I call attention to these exceptions to the generalizations not in any destructively critical attitude whatever. As I said, generalizations are very important and I know as well as any one present the difficulty of making a generalization to which there are no exceptions. I merely point out these exceptions as being important. They may mean a difference between profit and loss.

G. K. EGGLESTON [3]: Mr. Corson had a very proper idea in mind when he suggested yellow brass to conserve our tin supply but we had an idea along that line to reduce our metal costs. We tried it. Mr. Corson suggested one per cent lead. We left it up to three per cent, somewhere in there. On account of what happened, I didn't dare to go into the machine shop on account of the difficulties the boys had. When one speaks of easy machining in a jobbing shop and a high speed production job, it is highly different. As I recall, we had to cut our machine rate in half to handle that brass in comparison with the red brass we had been using.

CHAIRMAN ST. JOHN: I would like to add one remark to the early part of this discussion, the last sentence in Mr. Corson's paper, to which I think Mr. Clamer very properly took exception. As in any broad and sweeping statement, there are many exceptions to it. However, there is also a distinct germ of truth in Mr. Corson's statement in the sense that in any foundry during the time that standard procedures are being worked out, very frequent tests are necessary, tests on the quality of the metal produced, tests on the sand, tests on cores. These things must often be checked several times a day but as time goes on the necessity for frequency of tests becomes very much less.

I think that Mr. Corson is quite correct in what he probably means to say, that when the foundry procedure has been thoroughly established and everyone concerned with it has become familiar with the proper methods of carrying out his own particular duties, then these routine

[3] Metallurgist, Detroit Lubricator Company, Detroit.

tests can be made less and less frequently. Some of them possibly, can be abandoned altogether except in the case of an emergency. This, of course, doesn't apply to metal temperature testing. That can't be decreased. It doesn't apply to testing of raw materials for the foundry which must be purchased to specifications. It doesn't apply to tests which become necessary when emergencies arise in the foundry. But it is true that the frequency of tests necessary during the early stages in the improvement of foundry practice becomes less and less necessary as time goes on. I think Mr. Clamer will agree with that and I think it is worth emphasizing because some people might think these tests should be continued as a matter of form and safety, far beyond the limit of true necessity.

MR. CORSON: Mr. Kempf raised quite a number of points and I shall reply to them in the following order:

(1). *Iron in aluminum alloys.* It is most definitely established that iron is bad in all aluminum-silicon alloys. Just let them contain around one per cent iron and the big needles of the ternary constituent (iron-aluminum-silicon) will be clearly seen at quite a distance in all fractures. The strength of the metal will then drop dangerously. It will become porous and brittle.

Perhaps iron is not so harmful in aluminum-copper alloys, but I certainly shall argue against the statement that it makes the No. 12 alloy in its sand cast state more ductile. The latter never possesses an elongation above 2 per cent (a generous figure indeed), mostly 1 per cent, and with elongations so low the alloy should be considered as having practically no ductility. And if a few test bars containing iron happened to show 2 per cent elongation while control bars of a purer metal ran to 1.5 per cent elongation only, I should decline to be convinced. To be entitled to claim that iron increases the alloy's ductility we ought to have records of a long series of tests where every iron containing bar exhibited a substantial and regular margin of elongation. For example, a 4 per cent elongation in 100 test bars containing over 7 per cent iron, against 2 per cent in one hundred containing less than 0.5 per cent of it might form a convincing picture. In the absence of such records no claims of the kind should be made as it is not quite logical to expect iron to become a ductilizer by forming an additional amount of a brittle constituent between the grains or inside them.

(2). The complications I mention regarding aluminum silicon alloys are of a commercial nature. Scrap of straight 5 per cent silicon castings as well as that of No. 12 can be sold to be turned into secondary ingots. Scrap containing both copper and silicon cannot be so sold.

(3). Regarding magnesium and my statement concerning copper in the composition of some magnesium alloys. Of course, copper ought not to form a desirable component, but the book Engineering Alloys by Woldman and Dornblatt (1936) lists copper in some of the compositions mentioned. My information on that point came from this book.

(4). Regarding some special foundry aluminum alloys derived from the existing ones and not mentioned in my list, I know well enough that a number of proprietary alloys were built up around the old No. 12 alloy for instance. Small amounts of cerium, lithium, titanium, and so forth, were claimed to give the proprietary products certain highly desirable characteristics. The effects of most of these additions were ascribed to their potency as "deoxidizers." Needless to say, aluminum cannot be deoxidized nor does it need to be. No minor addition can successfully chase and destroy "films" of aluminum oxide on the metal's surface.

It is important to state further that in no case did the manufacturers of castings from such proprietary alloys exhibit an imposing table of the characteristics of their alloys as obtained in comparative tests performed simultaneously on a large number of control test pieces free from the particular addition. And while I would not go on record as stating that no addition might ameliorate an aluminum base casting alloy (used as cast), I shall state emphatically that the results from the use of most of these represent mere "sales talk."

Plastic Cupola Patching

By R. E. Wilke,* Waterloo, Iowa

Abstract

This paper describes the method used successfully since 1928, in the foundry, with which the author is connected, for patching the cupola melting zone. A stiff mixture of fire-clay and sharp sand is used in place of the conventional brick or block, resulting in economies in both refractory and labor.

Introduction

1. The usual practice in replacing the refractory material in the melting zone of a cupola, which has been burned away during a heat, is to use fire brick, cupola block, firestone, plastics, or a combination of these refractories. Another method of patching the melting zone consists of replacing the original lining with a stiff mixture of silica sand and fire clay, prepared in a pug mill.

2. After the slag and coke have been chipped out of the cupola, as shown in Fig. 1A the patching mud is applied by hand, starting at the tuyeres and working upward. Fig. 1B shows the melting zone partially patched, and Fig. 1C shows the complete patch after it has been allowed to dry. In Fig. 1D, the patch has been gone over to fill up the cracks and the well has been mudded. The cupola is ready to have the bottom doors raised and the sand bottom rammed in.

3. When possible it is desirable to patch the cupola in this manner the day previous to use as the mud will check during drying. Before raising the bottom doors and putting in the sand bottom, the mud patch is gone over and the cracks filled up. At the same time the blocks in the well are also covered with a thin layer of the patching material.

* Foundry Metallurgist, John Deere Tractor Co.

Note: Presented at a session on Refractories at the 43rd Annual A. F. A. Convention, Cincinnati, Ohio, May 16, 1939.

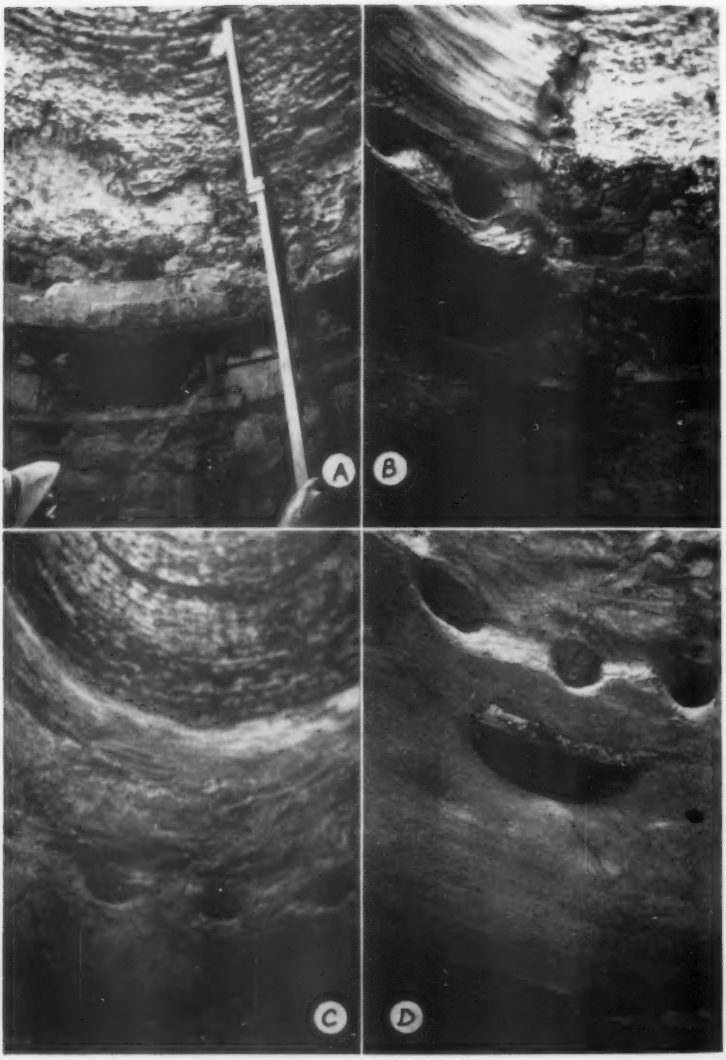


FIG. 1—VIEWS OF THE INSIDE OF A CUPOLA. (A) MELTING ZONE AFTER SLAG AND COKE HAVE BEEN CHIPPED OUT. (B) MELTING ZONE PARTIALLY PATCHED. (C) COMPLETE PATCH AFTER IT HAS BEEN ALLOWED TO DRY. (D) PATCH HAS BEEN GONE OVER TO FILL IN CRACKS AND THE WELL HAS BEEN MUDDED.

MATERIALS USED

4. A white silica sand and a fireclay both having a high fusion point are used to make up the patching material. The sand used is approximately 50 mesh and the fireclay, 10 to 12 mesh.

MIXING

5. Some type of a pug mill is recommended for properly mixing the sand, clay and water. Hand mixing will not give satisfactory results. Fig. 2 shows the type of pug mill used to mix and temper the patching material.

6. The sand and clay are shoveled into the mixer in the proportions used, approximately three shovels of sand to two of clay. Water is added continuously during the mixing. The mud is checked by "feel" as it comes out of the mixer for the proper consistency. Fireclay that is too fine will require more sand, more water and more intensive mixing. Too much clay will result in excessive cracking, spalling and is less refractory. Too much sand in the mix usually causes difficulty in getting the patching material to stick to the lining. The patching mud prepared in this ratio and properly tempered so that it will have the necessary "stickiness" and still not be too stiff to work, will have approximately 14 to 15 per cent moisture.

FIG. 2—PUG MILL USED TO MIX AND TEMPER THE PATCHING MATERIAL.

PATCHING THE MELTING ZONE

7. After the mud has been prepared, it may be used at once, as apparently no benefit is derived from allowing the mud to set. Two men, working from a "mud" board 2 ft. wide, which is inserted between the edges of the bottom doors and the bull ring, apply the patch to the melting zone. A third man wheels the mud and throws it up on the mud board.

8. Fig. 3 shows the lining used in the Whiting No. 10 cupola (96 in. shell) and the area to which the mud is applied. A layer of mud is applied to the new lining and after each heat the refractory burned away is replaced so that the original inside diameter is restored with the mud patch. This type of patch may be applied up to a thickness of 4 to 5 in. but if the lining is burned away to a greater extent, 1 in. split brick or 9 in. straights are embedded in the mud against the old lining.

9. Mud correctly mixed and applied will not slide off before drying. If the cupola lining is hot enough to cause steam to be formed when the mud is applied it must first be cooled by spraying with water as the patch will not stick if the lining is too hot. In chipping out the cupola, it is desirable to leave a rough surface to which to apply the patching material.

LINING LIFE

10. The melting zone of the cupola shown in Fig. 1A shows a maximum of 3 to 4 in. of the original block lining burned away

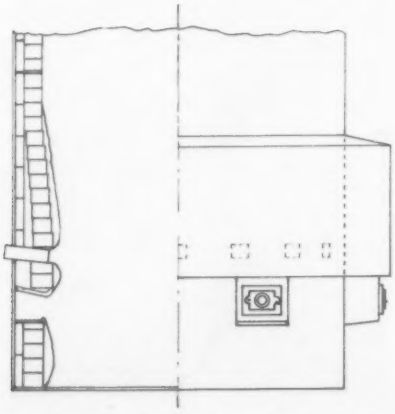


FIG. 3—INSIDE OF CUPOLA SHOWING THE AREA TO WHICH PATCHING MUD IS APPLIED.

after 1200 ton had been melted. No brick or block were used in the melting zone for patching during this time, the stiff mud alone being used to restore the lining to its original thickness. This particular cupola melted 130 to 150 tons in 7 and 8 hr. heats using a blast pressure of 16 to 20 oz.

11. The maximum heat taken from a cupola with this type of patching was 335 tons in 17.5 hr. with a blast pressure of 20 to 24 oz. Metal temperature at the spout ranges from 2820 to 2840°F. The longer heats with higher blast pressures are much more severe on linings regardless of the type of refractory used.

COST OF PATCHING

12. One ton of the prepared patching mixture having a ratio of 60 parts of sand to 40 parts of clay with the labor for mixing will cost approximately $4.00. The cupola completely patched as shown in Fig. 1D, will require from 2500 to 3000 lb. of the patching material at a cost of 5 to $6. Cost of the mud will vary depending on the geographical location of the particular foundry and the availability of sand and fireclay. When a good sharp sand is not close at hand, it may be more economical to reduce the ratio of sand to clay and use approximately a 50-50 mix, providing satisfactory refractoriness and service is obtained.

13. The economy resulting from this type of lining is due not only to the fact that a given amount of this refractory is much cheaper than an equal amount of other refractory but also due to the low labor cost of installation. Table 1 shows the time required for the various operations in patching the melting zone and preparing the cupola for the next heat.

Table 1

TIME REQUIRED FOR PREPARING A CUPOLA FOR THE NEXT HEAT

Chip out slag and coke	1 man — 20 to 40 min.
Patch melting zone	2 men — 45 min.
Mud well and fill cracks	1 man — 20 min.
Put in slag hole	1 man — 30 min.

REFRACTORY COST

14. A complete record of all sharp sand and fireclay for patching, and all block, brick, etc. for spouts, patching and relining used over a period during which over 66,000 tons of iron were melted, showed a refractory cost of $12.72 per 100 tons of melt.

DISCUSSION

Presiding: E. J. CARMODY, C. C. Kawin Company, Chicago, Ill.

CHAIRMAN CARMODY: Mr. Wilke, if you had used a finer mesh clay, could less clay be used with less cracking?

MR. WILKE: That might be possible, except when one uses a finer mesh clay he has a great deal more trouble mixing it. It would be harder to mix. Probably instead of one mixing, as we are doing now, it might mean a remixing. That has been our experience on finer clay.

CHAIRMAN CARMODY: How much finer clay did you use? Did you get as low as 100 mesh?

MR. WILKE: I can not say how fine it was, it was just the normal variation from one car of fire clay to another. If the clay runs finer, it requires a little better mixing than when coarser clay is used.

CHAIRMAN CARMODY: I have used a finer clay. You can use less of it, and the less clay used the fewer cracks that will appear on drying.

MEMBER: Is there any advantage in using a pug mill rather than a muller type mixer?

MR. WILKE: We have never tried out a muller. We have thought about it. In answering your question, I might tell you a little about the history of the development of the use of this lining. This happened before my time but I am told that originally when they tried to use this lining, they tried to mix it in a mortar box with a hoe, and it didn't work out very well. Then they thought that a little more intensive mixing would help them. So they mixed it in a box, like in the original method, but they put two or three men in there with rubber boots to give it kind of a kneading and mixing action, which did help it. Then they went to the use of the pug mill, which really did the work as far as mixing is concerned. I believe that a muller probably would do a better job of mixing, but it would cost more for labor because I doubt whether one could get the mixed material out of the discharge door. You would probably have to mix it in the muller and then shovel it out.

C. E. BALES:[1] This muller method of mixing would give a much better mud and it would be much easier to mix than in the pug mill. If you have a high percentage of non-plastic material, such as sand. it is almost impossible to get a good, thorough mix in a machine of the pug mill type. That type of mud should easily drop out of the bottom of the usual type foundry muller. We do it regularly on different types of plastic mixes. In that way it would be cheaper to mix and it should be a more effective mix. I would like to ask Mr. Wilke where he obtains the clay, and the sand he uses.

[1] Ironton Firebrick Co., Ironton, Ohio.

MR. WILKE: We use a northern Illinois clay, and the sand comes from northeastern Iowa. I think it is from a vein of the St. Peter sandstone.

MR. BALES: The author's cost figures on the materials shipped to his plant look very low. In other words, he has a freight item in there that would amount to quite a bit. Then the materials probably cost a couple of dollars a ton anyway, plus the freight, plus mixing. The $4.00 seem a little too low to me.

As I see it, there is one big disadvantage to this type of material as compared to cupola block, and that is, you must have plenty of cupola capacity. As Mr. Wilke mentioned, he prefers to line the cupola up the day before so it will be thoroughly dried out before it is put into service. In the automobile shops around Detroit, they have to get in the cupola at night, cool down the lining, tear out part of it, reline it and have the cupola in service the next morning. With this mud lining, with all the water that is in it, if one were to put the blast on when it is still wet, the steam would blow out the lining. I believe Mr. Wilke did mention he had tried it and it worked all right, but I'll say that was an unusual case.

MR. WILKE: There have been occasions where we have had to use the cupola the next day. We have four of the Whiting No. 10 cupolas, and we use two each day, alternating them. In that way we have an excellent opportunity of patching them on alternate days. But on several occasions we have had to use that same cupola on the following day and have done so very successfully.

CHAIRMAN CARMODY: Regarding the muller type mixer, I visited one foundry using this type of material to line their cupola. They were using a muller with a bottom discharge. In this case they allowed the material to set 24 hours. They covered it with wet bags for this period before using it in the cupola. They did not use sharp sand; they were using ordinary lake sand. And so far as they could determine, they were getting as good service from the lake sand as they had obtained with the sharp sand. In that particular shop the material appeared to get plastic during the operations. When the cupola bottom is dropped one can see the coke actually imbedded in the lining. I wondered if you had noticed that, Mr. Wilke?

MR. WILKE: We run up to as high as 330 tons to a heat over stretches of 16 and 17 hours, and occasionally have to use a fairly high blast along the latter part of the heat, although we would prefer not to, so I think it is reasonable to expect that lining does get more or less plastic. But the chances are that a lake sand would have more tendency to do that than a white silica sand.

MEMBER: How much steel scrap is contained in the melt in this particular cupola? What is the total carbon in the iron melted, and also how uniform is the carbon?

MR. WILKE: We use 10 per cent in our particular mix. We attempt to run about 3.25 total carbon. Of course uniformity of composition depends upon a lot of variables.

MEMBER: Can the author state positively whether or not the refractory contributes to the metallurgical efficiency or whether there is a decrease in efficiency in cupola operation? For several years I have used refractories other than fire brick, and I am satisfied myself personally that I am on the right track. But I may be wrong.

MR. WILKE: If the type of lining used interferes with the normal operation in any way, it might change the ultimate composition of the iron. Now so far as the effect of the composition of the lining itself, what effect that would have on the composition of your iron, I wouldn't attempt to answer that.

CHAIRMAN CARMODY: Is there anyone who cares to help out with information on their various types of linings, and whether they have any effect on the composition of the iron? That is, so far as the chemical composition is concerned, whether you have a pick-up from refractory materials. It is my belief, if you have good cupola operation you will not experience a pick-up or contamination from any of the common lining materials you might use. But then when one gets to defining good cupola operation that varies over a wide range according to the type of iron you are making.

MEMBER: My contention is that your cupola operation has more effect on the physical characteristics of the metal than any variation, that is, any slight variation, in the chemical analysis of the metal.

The Properties of Resin Bonded Cores

By Emile Pragoff, Jr.,* Wilmington, Del.

Abstract

Selecting a high melting point pine wood resin as typical of the improved resinous core binders now on the market, the author presents a series of studies on the more important properties of resin bonded cores and how these may be produced, varied and controlled in actual practice by an intelligent use of resinous binders.

The results obtained show that resinous binders increase in effectiveness the finer they are ground and, that, with proper mixing and baking at temperatures ranging from 300-500°F., strength and core hardness may be varied at will by regulating the binder concentration and moisture content of the core mix and the length of bake of the cores. Resinous binders are compatible with all of the commonly used binders and are particularly adaptable for use with sands of high clay content or with reclaimed sands. Even when mixed with moisture absorbing binders, cores bonded with resinous binders exhibit unusual resistance to loss of strength on storage.

A method for producing and testing one inch cubical core specimens by means of specially designed attachments to be applied to standard apparatus is described.

Graphs presenting the data are appended as well as photographs of the special testing equipment.

Introduction

1. Cores for foundry use are made today with a wide variety of binding materials. Linseed and other vegetable drying oils, core oils, cereals, resins, pitches, clays, and sulfite cellulose waste liquor find use in all types of foundries. The intelligent and economical use of these binders requires a knowledge of the factors

* Hercules Powder Company.

Note: This paper was presented at a session on Sand Research at the 43rd Annual A. F. A. Convention, Cincinnati, Ohio, May 17, 1939.

affecting their efficiency. Many investigators have studied the action of such materials as linseed oil,[†1] and core oils [2, 3, 4]. Cores bonded with resins require different technique in preparation and show somewhat different properties. The purpose of this study is to show the effect of different factors on the properties of resin bonded cores.

WHAT ARE RESINS?

2. To the practical foundryman the term "resin" will at once suggest rosin, a common constituent of core oils, and occasionally used itself in making a soft core. Resins are defined as amorphous vegetable products of secretion or disintegration, usually formed in special cavities of plants. The coniferous resins, or those obtained from the pine trees, are the most abundant resins occurring in this country. They are readily obtained by either of two methods. Over three hundred years ago the early settlers began to cut slashes in the pine trees in the South. The gum which flowed out was distilled to obtain turpentine and rosin. In 1910 H. T. Yaryan developed a process by which he produced rosin from the stumps of pine trees from the cut-over lands in Georgia and Mississippi. He found that by shredding pine stump wood, then extracting with gasoline and steam, he could obtain a slightly different resin and turpentine. This resin was called wood rosin, while the one obtained from the living tree was called gum rosin. These two rosins became known to foundrymen because of their wide use in core oil and to a lesser extent as dry binders. More recently there has been made available for use in foundries a new pine wood resin differing from either of these rosins. It has a higher fusion point than rosin, is almost entirely insoluble in gasoline and has the advantage over rosin in that it can be pulverized and stored without melting together to form lumps. It is thermoplastic or softens and flows on heating. This new resin has been widely used by foundries during the past four years as a dry binder. A study of its action under different conditions is given to represent the resin binders as a class.

FACTORS AFFECTING CORE PROPERTIES

3. Of the many factors affecting core properties, the following were studied:

† Superior numbers refer to Bibliography item numbers at end of this paper.

(1) Binder concentration.
(2) Degree of fineness of binder.
(3) Baking conditions—temperature and time.
(4) Moisture content of core mix.
(5) Mixing conditions.
(6) Addition of silica flour.
(7) Effect of other binders, such as cereals, oil, etc.
(8) Effect of reclaimed sand.
(9) Storage of baked cores prior to use.

4. In studying each of these factors we have endeavored to control the others closely, varying the conditions one at a time to observe the effect upon:

(a) Green compression strength.
(b) Green and dry permeability.
(c) Dry tensile strength.
(d) Dry transverse strength.
(e) Hardness.

TESTING PROCEDURE

5. A test procedure was worked out which was adapted to the laboratory equipment used in the investigation. It deviates only slightly from the tentative standard procedure of the A.F.A. Subcommittee on core tests.**

6. In making core tests the following procedure was employed: 3,000 grams of air-dried sand was placed in a laboratory size muller type mixer. Dry binders were weighed into a tared beaker and added to the dry sand. Two minutes dry mixing was found to be sufficient with the sands used. Tempering water was then added and mixing continued for 3 min. A total mixing time of 5 min. was established for use in all tests. After mixing, the core mix was discharged from the mixer into a 2-gal. pail fitted with a friction top. Moisture determinations were made upon a portion of the mix using the Dietert moisture teller.

7. The test core specimens were rammed with a standard rammer, using suitable rammer heads to prepare six tensile, three transverse and four compression core specimens.

** Testing and Grading Foundry Sands and Clays, 4th Edition, 1938.

8. When this research was begun, there existed no instrument capable of measuring compression strength up to 1,000 lb. per sq. in. which was approved by the A.F.A. Committee on Foundry Sand Research. The Dietert Universal sand strength machine is designed to measure compression strength only up to 300 lb. per sq. in. We found that by using a 1-in. cube specimen instead of the 2-in. cylindrical specimen we could increase the effective load on the specimen. Since the surface of the 2-in. cylinder was 3.1416 sq. in. and that of a 1-in. cube was 1.0 sq. in., we have increased the range from 300 to 300 x 3.14 or 942 lb. per sq. in.

9. The 1-in. cube specimens were prepared by weighing approximately 28 grams of core mix into the special core box assembly shown in Fig. 1, and ramming with three blows of the standard rammer. After baking, the specimens were broken on the Dietert sand strength machine, adapted as shown in Fig. 2. To do this we made two brass discs 2 in. in diameter and 0.5 in. thick to hold the specimens. To these were fastened supports in such a way that the specimen was held firmly with the center of the 1-in. core at the center of the 2-in. specimen holder. This procedure worked out nicely and we obtained good breaks on all specimens as shown by the pyramidal shape of the upper and lower halves of the broken specimens. Good reproductibility of results was obtained.

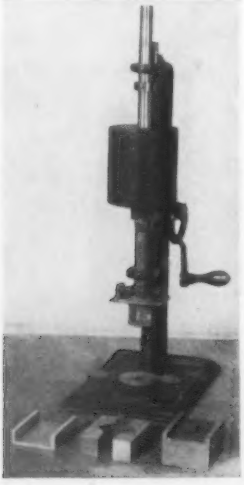


FIG. 1—SAND RAMMER AND ONE INCH CUBE BOX ASSEMBLY.

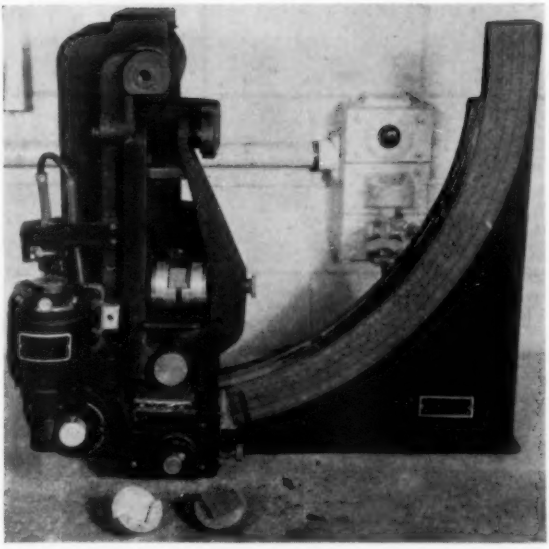


FIG. 2—SAND STRENGTH MACHINE ADAPTED FOR MEASURING DRY COMPRESSION STRENGTH OF ONE INCH CORE SPECIMENS.

10. Cores were baked under the desired conditions of temperature and time in an electrically heated and controlled oven. This oven, which was made by Young Brothers Company, Detroit, Michigan, was not equipped with the customary air blower and rotating shelf of the A.F.A. standard oven, but was found to give reproducible results consistent with those obtainable in the foundry. After baking, cores were allowed to cool for one hour and then broken in the usual fashion on a Dietert Universal sand strength machine. Where the individual specimens differed by more than 15 per cent from the average of a series, the results were considered not sufficiently accurate and a new core mix was prepared and baked. The large number of tests made obviated the possibility of using the A.F.A. 50 to 70 standard sand throughout. It was used only in the tests on storage of baked cores. In all other tests the sand used was a New Jersey No. 80 dry silica sand furnished by the Dodge Steel Company of Philadelphia, Pennsylvania. The characteristics of this sand are illustrated in Table 1. It is a subangular sand of 71 fineness number, used widely in core work in Eastern foundries. The reclaimed sand used in one series

of tests was obtained from a large steel foundry and is considered representative. The screen analyses of these sands are in Table 1.

Table 1

Sieve	*New Jersey Silica*	*Reclaimed Sand*
6	.....	....
12	.....	0.5
20	.....	2.9
30	.....	9.6
40	Trace	19.8
50	5.9	23.3
70	34.3	18.3
100	35.3	9.1
140	18.2	4.1
200	5.6	1.7
270	0.5	0.3
Pan	Trace	1.6
Per Cent Clay	.....	8.2
Total	99.8	99.3
Fineness No.	71	51

MATERIALS USED

11. The pine wood resin used in this investigation was taken from a bag representing material as used commercially. The specifications on this material were as follows:

100 per cent passing 30 mesh
90 per cent passing 80 mesh
60-80 per cent passing 200 mesh

Unless otherwise specified, it is this resin that is referred to whenever the word resin is used in this text.

EFFECT OF BINDER CONCENTRATION ON CORE PROPERTIES

12. The effect of concentration of binder in a resin bonded core is illustrated in Fig. 3. The hardness increases in a straight line function, while tensile strength and compression strength increase in parabolic function. The permeability of a core is affected very little by the concentration of resin binder. This curve is definitely in line with the findings of other investigators who have studied linseed oil and core oils. The data in Fig. 3 indicate that a wide range of core strengths and hardness may be obtained by the use of resin binders.

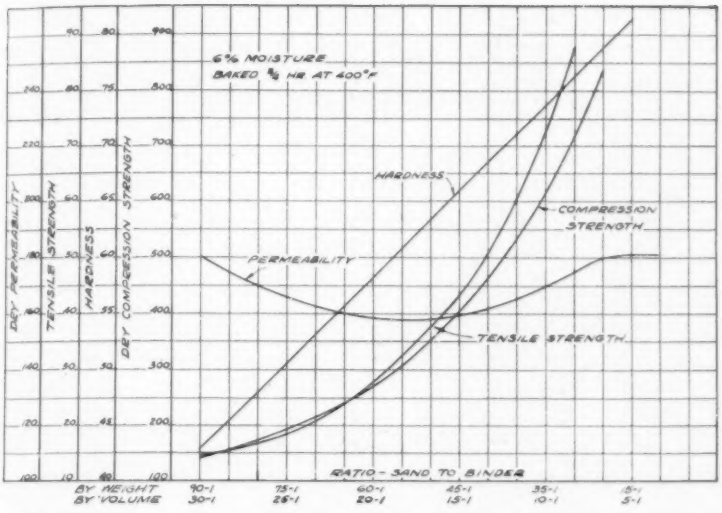


FIG. 8—EFFECT OF BINDER CONCENTRATION ON PROPERTIES OF RESIN BONDED CORES.

EFFECT OF FINENESS OF BINDER

13. In order to establish the optimum fineness to which a resin should be pulverized, a series of tests was made in which the solid resin was pulverized, screened and separated into various fractions which were used in preparing a series of cores. The pulverized material so prepared was separated into the following fractions shown in Table 2.

Table 2

Number	*Mesh*	*Passing Mesh*	*Held on Mesh*
1	30	30	50
2	50	50	70
3	70	70	100
4	100	100	140
5	140	140	200
6	200	200	270
7	270	270	Pan
8	100+	100	

14. These fractions were tested for efficiency as core binders in the following core mix:

New Jersey Silica Sand.....3000 grams
Resin Fraction..... 67 grams
Moisture ..... 6 per cent

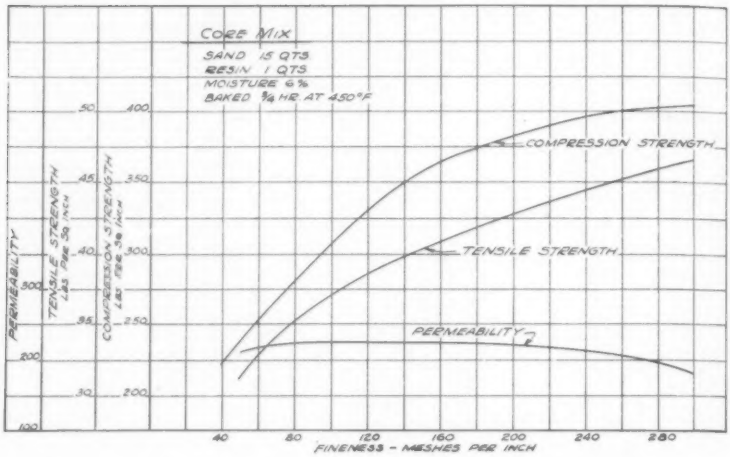


FIG. 4—EFFECT OF FINENESS OF BINDER ON PROPERTIES OF RESIN BONDED CORES.

15. Fig. 4 gives data on the core properties obtained with the various fractions. From visual examination of the baked cores, it was apparent that the coarser fractions, numbers 1 and 2 of Table 2, gave mottled cores in which particles of baked binder stood out clearly against the lighter colored sand. Tensile strength rises steadily from 31 lb. per sq. in. using 30 mesh binder to 46 lb. per sq. in. using 270 mesh material. Transverse strength is similarly affected. Dry compression strength reaches its maximum at 270 mesh. Green bond is also best in the finer fractions. Green permeability drops only slightly in the finer fractions, being regulated by the sand fineness. Dry permeability is similarly affected. Hardness is fairly constant at 65, except with the coarser fractions.

EFFECT OF BAKING CONDITIONS

Baking Temperature

16. Fig. 5 shows the effect of baking temperature on dry compression strength of resin bonded cores. A baking time of 1 hr. was used and the temperature was varied from 300-600°F. The compression strength rises as the baking temperature is increased, reaching a maximum at 500°F. and dropping rapidly thereafter. It is probable, that the bond developed by the resin is completely destroyed at temperatures above 700°F. From our observations, we have learned that resin bonded cores develop strength in an

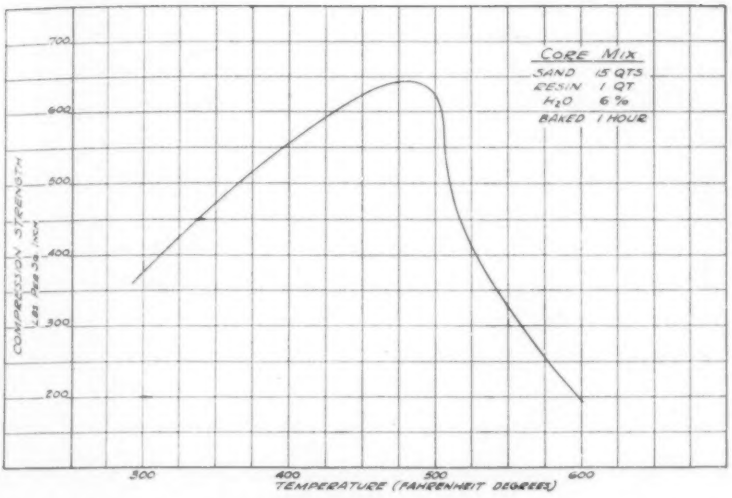


FIG. 5—EFFECT OF BAKING TEMPERATURE ON DRY COMPRESSION STRENGTH OF RESIN BONDED CORES.

entirely different manner from oil bonded cores. Resin bonded cores do not require the presence of atmospheric oxygen during baking. The stages which we would find in the baking of a resin bonded core would be as follows:

(1) The evaporation of moisture from the core.

(2) A gradual rise in temperature of the core which would cause the binder to melt and flow between the grains.

(3) The rising temperature would drive off from the resin volatile material and the resin would harden. Thus, we can understand the reason for higher compression strength as the temperature increases up to 500°F.

17. In the baking of an oil bonded core, we have similar stages where the moisture is first driven off. Next the temperature of the core is raised, and volatile matter is driven out of the binder. Finally, we have a rapid development of bond in the core which is caused by the oxidation or polymerization of the oil. The increased demand for speed in core baking indicates a more extended use of resin binders. Our laboratory observations on the baking of resin bonded cores have been proved true in the foundry. Many foundries prefer baking conditions of ¾ hr. at 450°F. per in. of core thickness. An advantage of baking at the maximum

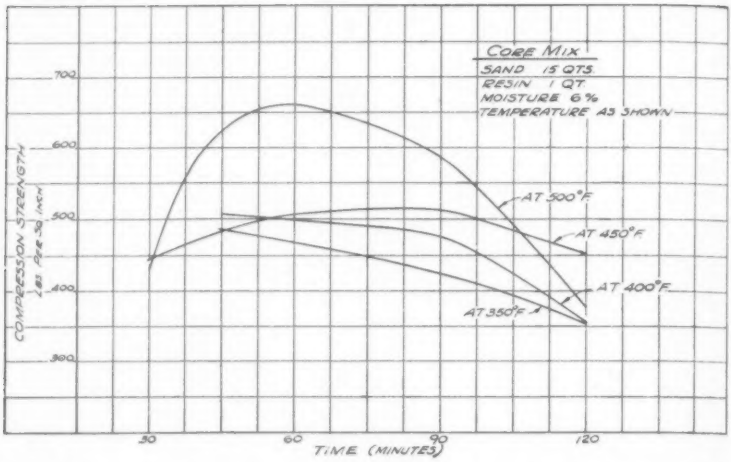


FIG. 6—EFFECT OF BAKING TIME AT VARIOUS TEMPERATURES ON DRY COMPRESSION STRENGTH OF RESIN BONDED CORES.

practical temperature would be that the cores would show less gas when burning out than those cores which had been baked at 350°F.

Baking Time

18. Fig. 6 shows the effect of baking time. We see that resin bonded cores are not extremely sensitive to extended baking periods and cores of good compression strength may be obtained by baking from 30 to 90 min. at 400 or 450°F. At 500°F., the curve reaches a peak in 45 min. and declines gradually thereafter. These data were determined on a core mix containing 6 per cent moisture. We will next show what happens when the moisture content is varied.

EFFECT OF MOISTURE CONTENT

19. Fig. 7 gives data showing the effect of moisture content on core properties of resin bonded cores baked 45 min. at 450°F. As the moisture content of the mix is increased, the compression strength and hardness improves, leveling off above 6 per cent. On the other hand, tensile strength drops rapidly between 3 and 6 per cent and levels off thereafter. Transverse strength reaches a maximum at 6 per cent. These data show interesting contrast to the data presented by Dietert on the effect of moisture in oil-sand cores. One explanation of the improved compression strength is

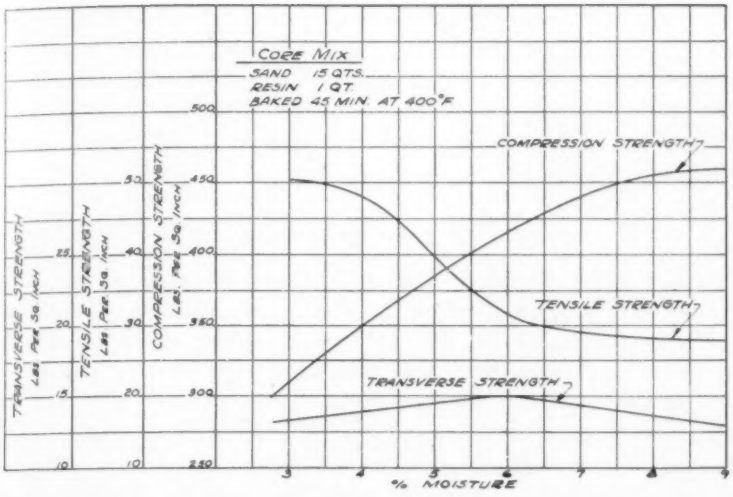


FIG. 7—EFFECT OF MOISTURE CONTENT ON THE PROPERTIES OF RESIN BONDED CORES.

that the higher moisture reduces the effective baking time at the maximum temperature.

20. Since compression strength and hardness are more dependent upon the surface condition of the core, we naturally expect these properties to be more affected by moisture than tensile and transverse strengths. These latter properties are more dependent upon the conditions existing in the interior of the core. There is the other possibility of partial migration of resin binder to the surface of the core. Results obtained in daily use show that this supposed migration does not result in cores having soft centers. These data indicate the necessity of close control of moisture content in core mixes.

EFFECT OF MIXING CONDITIONS

21. The effective use of any binder requires a knowledge of the preferred mixing practice. We have seen arguments pro and con on the question of how oil-sand mixes should be prepared. In our tests, the resin binder was added to the dry sand and mixed dry for 2 min. The tempering water was then added and the wet mixing of an additional 3 min. followed. A total mixing time of 5 min. was shown to be optimum for the laboratory size muller type mixer. A longer mixing period than 5 min. had the tendency to dry out the mix. As we have just shown, a lower moisture con-

tent will give lower strength cores. In order to answer any questions which might arise, the practice of adding resin binders to wet sand has been studied and we find that when the sand is not too wet, there is no mixing problem. Usually, slightly lower strengths will be obtained when it is necessary to add resin binder to wet sand.

EFFECT OF STORAGE OF BAKED CORES

22. Some foundries find it necessary to prepare their cores in advance and to store them for several days prior to use. Under such conditions, especially in humid weather, it is desirabe to select core making materials such that the cores will not lose strength on storage. One of the inherent natural properties of resin binders is that they are not water absorbent. Hence, cores prepared from them do not become weaker on storage. Table 3 shows that a set of cores prepared from silica sand bonded with resin maintained the same core strength over a storage period of 30 days. The initial compression strength of these cores was 400 lb. per sq. in. Twenty-four hours later cores from the same set showed 422 lb. per sq. in. After intervals of 7, 15 and 30 days, we find the strength and hardness relatively unchanged.

Table 3

EFFECT OF STORAGE ON STRENGTH OF RESIN BONDED CORES

Storage Time	1 Hr.	24 Hrs.	24 Hrs.*	7 Days	15 Days	30 Days
Tensile Strength (lb. per sq. in.)	36	36	33	36	36	37
Compression Strength (lb. per sq. in.)	400	422	404	432	434	432
Hardness	65	65	65	65	65	65

Core Mix.....Silica Sand—15 qts., Resin Binder—1 qt., Moisture—6 per cent—Baked ¾ hr. at 400°F.

* Denotes properties after storage in sealed container at 100 per cent relative humidity. All other cores were allowed to stand under normal atmospheric conditions. In the tests, 24 cores were prepared and 4 were broken at each interval.

A COMPARISON OF SEVERAL RESINS

23. In order to show the difference between the improved pine wood resin binders of today and the rosins which have been used for so many years, the data in Table 4 have been obtained. Here are shown the data on the compression strength obtained with three representative types of coniferous resins and the pine wood resin used in this investigation.

Table 4

COMPARATIVE DATA ON CONIFEROUS RESINS

	No. 1 Rosin	*No. 2 Rosin*	*No. 3 Rosin*	*High Melting Pine Wood Resin*
Green Bond	1.1	1.1	1.05	1.05
Green Permeability	90.0	90.0	95.0	90.0
Dry Permeability	160.0	180.0	170.0	170.0
Dry Tensile Strength (lb. per sq. in.)	28.0	28.0	29.0	36.0
Dry Transverse Strength (lb. per sq. in.)	10.5	15.0	12.5	17.0
Dry Compression Strength (lb. per sq. in.)	194.0	208.0	273.0	446.0
Hardness	55.0	60.0	60.0	65.0

Core Mix.....Silica Sand—15 qts., Pulverized Resin—1 qt., Water —6 per cent—Baked ¾ hr. at 400°F.

EFFECT OF DIFFERENT SANDS

24. Resin bonded cores are made today with a wide variety of core sands. When they are made with washed silica sands, it is common practice, especially in steel foundries, to add small amounts of fireclays, bentonite, or cereals since these materials assist in the preparation of a better core. Natural bonded sands are used effectively with resin binders. It is interesting to note that core sand mixes containing clay do not absorb resin binders in the same fashion that they would oil binders. We have not endeavored to show the effect of sand fineness upon the properties of resin bonded cores but would expect it to follow more or less the same line as with other binders. Probably there is an optimum sand fineness which would show the greatest compression strength and then as the fineness increased, the strength would be lowered.

EFFECT OF RECLAIMED SANDS

25. Resin binders are particularly adapted to use with reclaimed sand. Many large steel foundries find it both economically practical and highly desirable to use their reclaimed sand in core mixes. When mixtures of new and reclaimed sand are bonded with a resin binder, they produce strong, hard but readily collapsible cores. In Fig. 8 are given the data showing the effect of reclaimed sand. Starting with 100 per cent silica sand, we have replaced

20, 40, 60, 80 and 100 per cent with reclaimed sand. Keeping the binder ratio constant and using a moisture content of 6 per cent, we find the presence of reclaimed sand has increased compression strength, both green and dry, and hardness, without too seriously reducing the permeability. Tensile strength is not particularly affected by the presence of reclaimed sand. (See Fig. 12 in Addenda).

EFFECT OF ADDED MATERIALS

26. In preparing cores having definite desired properties, resin binders may be used in combination with other common ingredients of core mixes such as silica flour, fireclays, bentonite, cereal binders, linseed oil and core oils. In Fig. 9 the effect of silica flour is illustrated. Starting with no silica flour and adding 5, 10 and 15 per cent, we obtained a marked increase in compression strength and hardness and a marked reduction in permeability. Tensile strength was affected very little. The collapsibility of such cores would undoubtedly be affected. The less open cores containing silica flour should collapse less readily because of their lower permeability. The resin would break down readily from the heat during pouring, but the silica would resist to some degree the contraction of the metal.

27. The presence of bentonite and fireclays increases the efficiency of resin binders as mentioned previously. When we

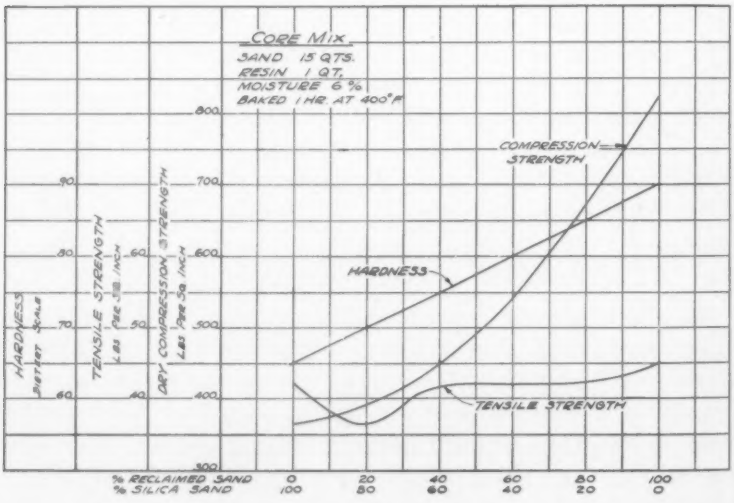


FIG. 8—EFFECT OF RECLAIMED SANDS ON THE PROPERTIES OF RESIN BONDED CORES.

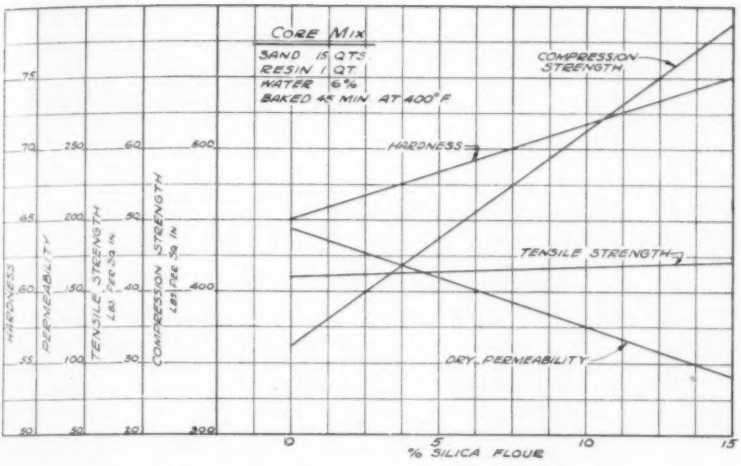


FIG. 9—EFFECT OF ADDED MATERIALS ON THE PROPERTIES OF RESIN BONDED CORES—EFFECT OF SILICA FLOUR.

added clay to resin bonded core mixes, we obtained improved compression strength and hardness, better green bond, and only slightly lower permeability.

28. The cereal binders, particularly those which are derived from gelatinized starches are widely used in core mixes for two purposes; first, to produce green bond and second, to give some dry bond. Since the resins are not water soluble, their effect as green bond producers might seem to be an anomaly. In Fig. 10, it will be seen that resins do produce appreciable green compression strength. In this, they are not as much affected by moisture content as are cereals and clays. Resins work well in combination with cereals and Fig. 10 shows that a mixture of the two gives considerably better dry compression strength than either alone. This fact has already been the means of considerable savings to foundrymen. Resin binders, by virtue of their moisture resistance, are valuable for use in improving the moisture resistance of cereal bonded cores.

29. Linseed oil alone or in core oils is probably the most widely used core binder. Resin binders can be used effectively and economically in conjunction with drying oils. The resin and oil on heating together develop a stronger bond which apparently is developed more rapidly in baking.

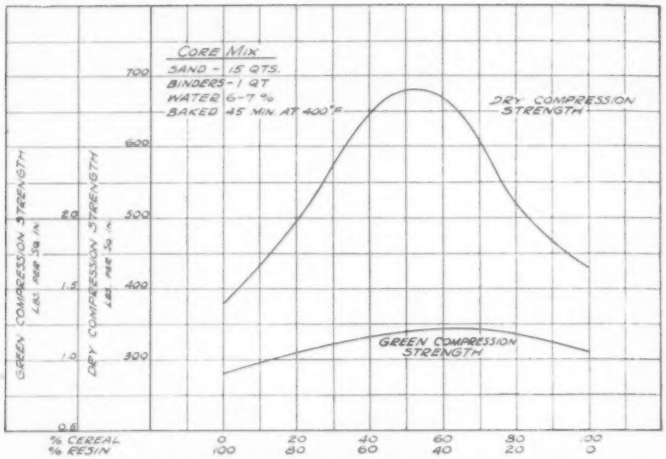


FIG. 10—EFFECT OF ADDED MATERIALS ON THE PROPERTIES OF RESIN BONDED CORES—EFFECT OF CEREAL BINDER.

THE REPRODUCIBILITY OF RESULTS

30. Since we are obliged to use a test procedure which differed from the A.F.A. standard, the question naturally arises regarding the agreement between results obtained using a one inch cube specimen and a two inch cylindrical specimen. In Table 5 we have presented data obtained on five core mixes tested following each procedure. A mix was selected which would give a dry compression strength below 300 lb. per sq. in. The respective core specimens were baked under identical conditions. The data in this table are the averages of four determinations on each type of specimen. Using a 1 in. cube specimen, we obtained a dry compression strength of 249 lb. per sq. in. as compared to an average of 260 lb. per sq. in. for the 2 in. cylinder. Thus, a difference of 11 lb. per sq. in. or about 4 per cent. The individual sets deviate from the average very little, indicating good reproducibility of results.

Table 5

COMPARATIVE DATA ON DRY COMPRESSION STRENGTH OF CORES OBTAINED WITH TWO INCH CYLINDRICAL AND ONE INCH CUBE SPECIMENS.

	1	*2*	*3*	*4*	*5*	*Average*
2 in. Cylinder	263	264	264	242	270	260
1 in. Cube	259	229	265	254	238	249

Core Mix.....Sand—100 qts., Resin Binder—4 qts., Bentonite—1 qt., Water—6 per cent—Baked ¾ hr. at 450°F.

31. The deviation of the average results using 1 in. cube from the results by the standard A.F.A. method is only 4.4 per cent. The mean variation of any set of data from the average for that method is small, about 2.5 per cent for the 2 in. cylinder and 5.0 per cent for the 1 in. cube.

CONCLUSIONS

32. The conclusions of our experiments on cores bonded with high melting point pine wood resin are as follows:

(1) Resin bonded cores require different treatment in preparation and possess unusual properties of value to foundrymen.

(2) The concentration of resin binders in the core mix determines the strength and hardness which will be obtained.

(3) The degree of fineness of a resin binder affects its efficiency as a binder.

(4) Resin bonded cores may be baked satisfactorily at temperatures from 300-500°F.

(5) Resin bonded cores may be baked for various periods from 30 to 90 min. per in. of core thickness without serious loss of strength.

(6) Resin bonded cores possess extremely high collapsibility. This is illustrated by their breakdown above 600°F. Resin bonded cores are particularly adapted to use in the production of high shrinkage castings.

(7) In resin bonded cores, increasing the moisture up to 6 per cent or 7 per cent improves the compression strength and hardness. Tensile strength is differently affected, decreasing as the moisture is raised.

(8) The preferred mixing practice for core mixes containing resin binders is to add the dry pulverized resin to dry sand, mixing well then tempering with the necessary water.

(9) Resin binders are compatible with all of the other commonly used binders.

(10) Resin binders are particularly adaptable to use in mixes containing sands having a high clay content.

(11) The use of resin binders will make it possible for many foundries to utilize their reclaimed sand. Such cores are adapted to many jobs, now done only with new sand.

(12) Resin bonded cores are extremely resistant to loss of strength on storage.

(13) Resin bonded cores are a means of reproducing daily and controlling within narrow limits desired core properties of hardness and compression strength.

ACKNOWLEDGMENT

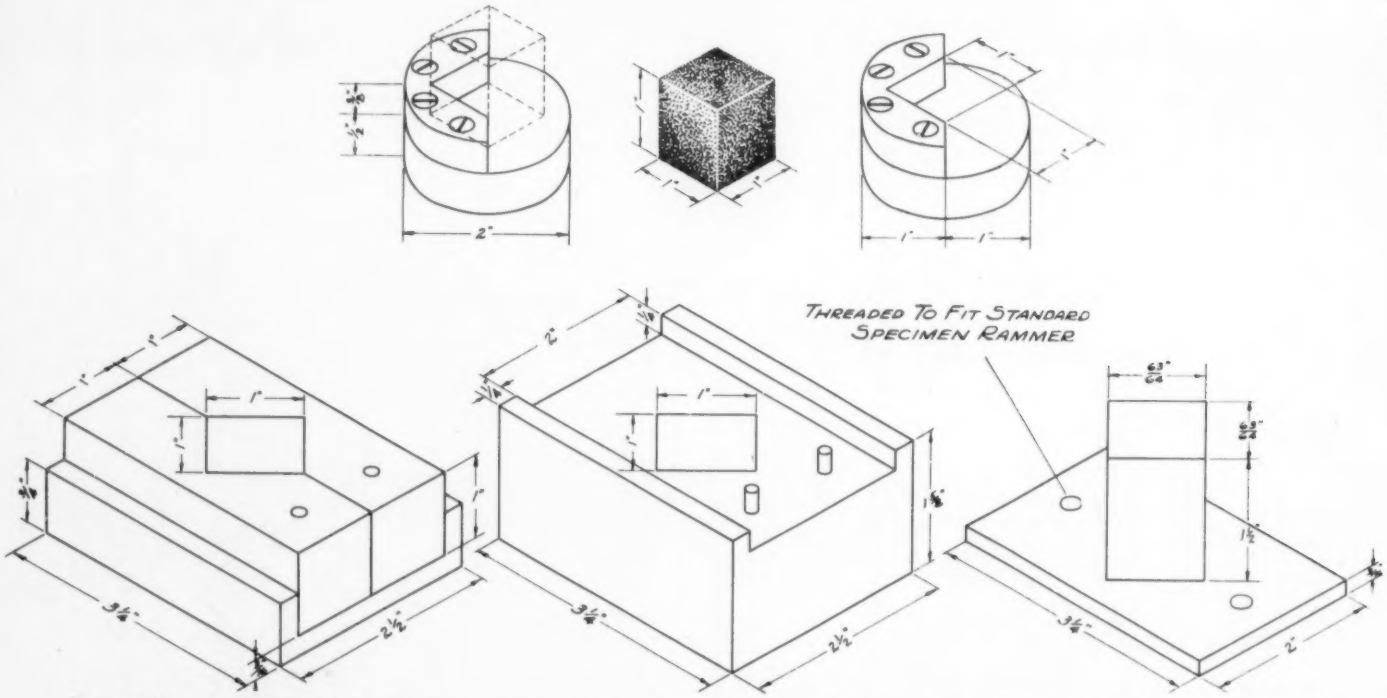
33. The author is deeply indebted to F. F. Dore, American Steel Foundries, Verona, Pa., and to E. C. Troy, Dodge Steel Co., Philadelphia, Pa., for suggestions in this work. These steel foundrymen guided us in selecting the proper range of conditions and otherwise assisted us in the research. The silica sand was furnished by Dodge Steel Co.

Addenda

34. A detailed drawing of the core box assembly for preparing and adapter for breaking one-inch cube dry compression core specimens is shown in Fig. 11.

Effect of Reclaim Sand

35. The data in Fig. 8 show that tensile strength was not greatly affected by the use of reclaimed sand, while compression strength was. This is believed to be caused by the spread in grain fineness of the respective sands used in these tests. The reclaimed sand and new silica sand had grain fineness numbers of 51 and 71, respectively. To test the action of a reclaimed sand having grain fineness nearly equal to that of the new silica sand, a further set of data were obtained, as illustrated in Fig. 12. Here the reclaimed and new sands were from the same origin; the reclaimed sand represented a synthetic bonded molding sand. It had a grain fineness number of 69, while that of the new sand was 71. Here the tensile and compression strengths appear to be equally affected as more reclaimed sand is used in the mix. Permeability is less affected since the two sands do not exhibit any packing. Green compression strength is considerably affected, reaching 4.6 lb. per sq. in. when using 100 per cent reclaimed sand. These data serve to illustrate the opportunities which are afforded to many foundries in the use of reclaimed sand in cores.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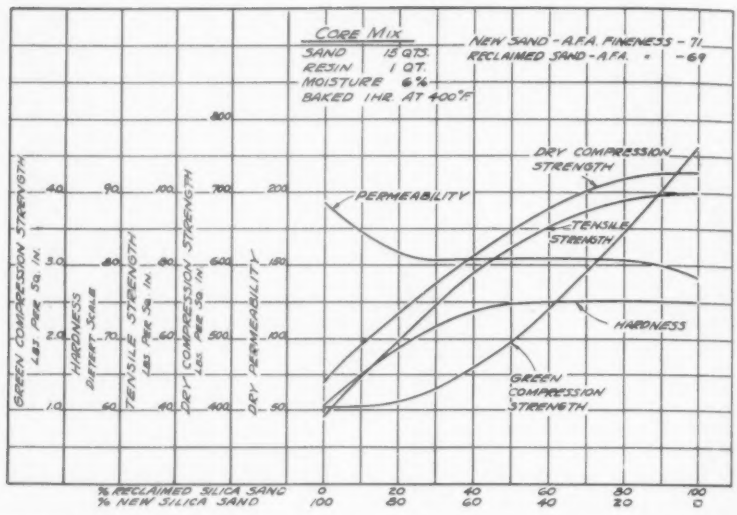

FIG. 12—EFFECT OF RECLAIMED SAND ON PROPERTIES OF RESIN BONDED CORES. RECLAIMED SAND OF GRAIN FINENESS 69 AND NEW SILICA SAND OF GRAIN FINENESS 71 WERE USED.

Effect of Other Binders:

36. Paragraph 29 states that resin binders can be used effectively and economically in conjunction with drying oils. The data in Table 6 serves to illustrate the effect of using pulverized resin binder with linseed oil.

Table 6

EFFECT OF USING PULVERIZED RESIN BINDER WITH LINSEED OIL

		No. 1	*No. 2*	*No. 3*
50-70 A. F. A. Std. Sand	Lb.....	1,000	1,000	1,000
Linseed Oil	Lb.....	10	7.5	5
Resin Binder	Lb.....	....	2.5	5
Oil Replaced, Per cent	.....	0	25	50
Baking Time, Hr.	.....	1½	1½	1½
Tensile Strength at 350°F	.....	66	104	76
(Lb. Per Sq. In.) 400°F	.....	147	152	145
450°F	.....	99	112	112
Transverse Strength at 350°F	.....	20	25	14
(Lb. Per Sq. In.) 400°F	.....	38	35	35
450°F	.....	25	32	29

The weight ratio in the initial mix, 1 to 100, is about the same as 1 to 60 by volume.

Bibliography

1—Campbell, H. L., *"Baking Practice in Oil Sand Cores."* Trans. A.F.A. Vol. 37 (1929) pp. 289-294.

2—Dietert, H. W. and Woodliff, Earl, *"A Study of Core Hardness."* Trans. A.F.A., Vol. 45 (1937) pp. 545-558.

3—Zirzow, E. C., *"Sand Control in a Malleable Foundry."* Trans. A.F.A., Vol. 45 (1937) pp. 134-155.

4—Fuerst, Chas. *"Sand Control in Relation to Steel Foundry Production."* Trans. A.F.A., Vol. 45 (1937) pp. 325-331.

5—Willsea, Jasper, *"What Happens to a Core in the Core Oven."* Trans. A.F.A., Vol. 46 (1938) pp. 739-757.

DISCUSSION

Presiding: H. S. WASHBURN, Plainville Casting Co., Plainville, N. J.

MEMBER: What type of cereal binder was used in the tests illustrated in Fig. 10?

MR. PRAGOFF: The mogul type.

MEMBER: The paper indicates best results are obtained with resin binder at moisture contents of 6 per cent. When using blowing machines, we cannot operate at over 3 per cent moisture. Will such a binder cause sticking in blowing machines?

MR. PRAGOFF: In oil-cereal mixes, at low moisture content, resin binders have been used as partial or complete oil replacement. In such cases, sticking may be prevented by careful control of moisture and addition of fuel oil to give the oily feel desirable in such mixes. Mixes containing resin binder can be prepared to possess good flowability.

MEMBER: Has the author any data on collapsibility?

MR. PRAGOFF: The baking curve in Fig. 5 illustrates the breakdown of the resin bonded cores as temperatures over 500° F are reached. We have not yet obtained high temperature collapsibility data, but hope to have more along this line later.

MEMBER: Has the author any data on the evolution of gas from resin bonded cores?

MR. PRAGOFF: Our laboratory has not yet obtained data on this question. However, we have the report of one foundry control laboratory which showed that cores bonded with resin and core oil showed less gas evolution than straight oil bonded cores.

CHAIRMAN WASHBURN: This paper is interesting in its presentation of data on binders which can be used with sands containing clay. We have usually considered that we must wash core sands free of clay. This new approach should be of interest to many foundries.

Castings for the British Gas-Industry†

By F. Whitehouse,* Sheffield, England

Abstract

This paper gives an idea of the casting conditions encountered in England. The author describes in general the use of castings in the gas industry. The methods of manufacture of retorts, coke oven castings and gas purifier boxes are detailed. Data on sand, and physical tests, etc., are included. Photographs illustrate the various parts discussed.

Introduction

1. The gas industry, it has been aptly said, was born in a cast iron retort. The cave shown in Fig. 1 was, perhaps, fittingly, the "laboratory" in which Murdoch experimented in making coal gas in his mother's iron teapot and iron kettle.

2. Descriptions of the retorts used when the great adventure of lighting Westminster with gas was entered upon differ. In the "Day Books" of the author's firm, there are despatch entries of retorts sent to the original Westminster Gas Light & Coke Co., though without details of size and shape. From then to the present day, the records reflect all phases of gas-production plants. It is of the foundry's connection with these that the author is to write.

3. Fig. 2 is a drawing of a typical small gas plant installation of 1866. It might have been used in small villages or on the estates of people such as the Duke of Devonshire or Earl Fitzwilliam, who had gas plants of their own. The type of retort used is shown in Fig. 3A, though it should be said that the shape was often a plain "D" rather than a hump back variety as illustrated here. Round retorts also were used.

4. The retorts were cast vertically with the flange at the bottom. The core box, in the case of "D" shapes, was a cast iron cradle of the correct shape. The procedure was to wrap straw rope round

†Official Exchange Paper, Presented on Behalf of The Institute of British Foundrymen.

*Manager of Foundry and Pattern Shop, Newton, Chambers & Co., Thorncliff Iron Works.

Note: Presented at Gray Iron Session of the 43rd Annual A.F.A. Convention, Cincinnati, Ohio, May 16, 1939.

FIG. 1—INSIDE OF CAVE WHERE COAL GAS WAS FIRST MADE—CAVE WHERE MURDOCH BURNED CANNEL COAL—SITUATED IMMEDIATELY BELOW ENTRANCE TO BELLO MILL.

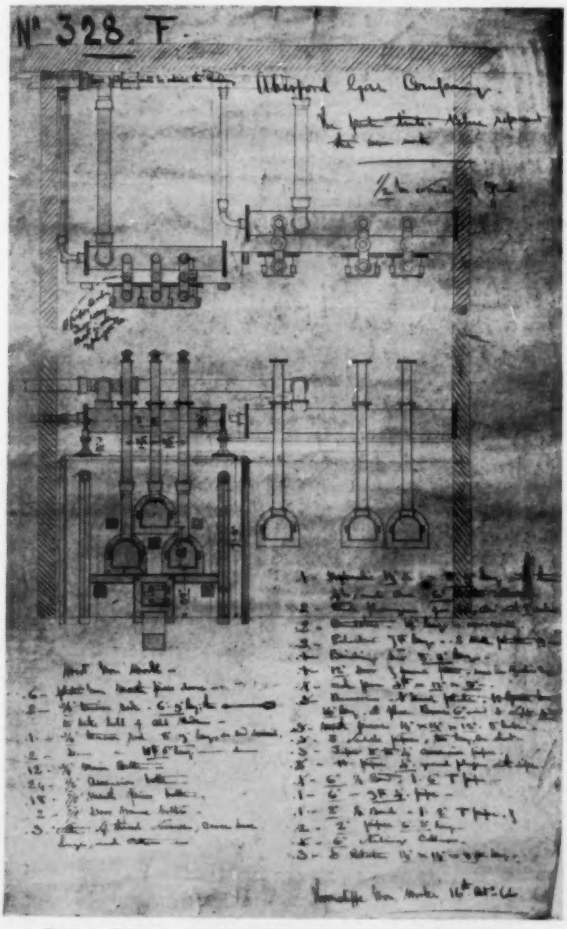

FIG. 2—DRAWING OF GAS-WORKS INSTALLATION OF 1866.

a "D" shaped barrel, line the cradle with loam, bed the wrapped barrel in, and make the top shape by means of a wood strickle. The whole was then put to dry and after drying the core was taken out of the cradle and blacked. The iron used for these retorts was phosphoric, of the cold-blast variety and made from local ironstone.

5. Fig. 3B is a cast iron pattern for a retort mouthpiece which is still in use today.

6. The retort, which nowadays must withstand very high temperatures, is no longer made of cast iron. The mouthpiece is, as it were, an external extension to a silica brick retort. The metal of which these mouthpieces are made, now is varied to suit operating circumstances. This will be explained in another part of the text.

7. All castings are made in green sand. The point of the three openings in the molding box, only one branch on the pattern, takes up different positions on the mouthpiece. The box is meant to accommodate them all thus avoiding a great variety of boxes. The apertures not in use are covered with plates. The core barrel and strickle, (A, Fig. 3B), is used to make the core for the socket

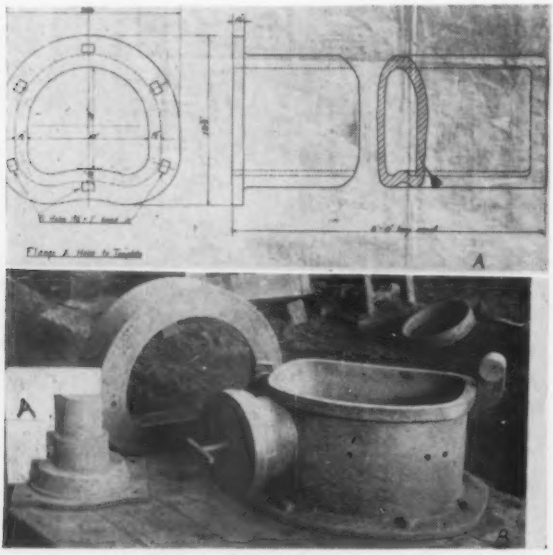

FIG. 3—(A) A TYPE OF "D" RETORT MADE IN CAST IRON, (B) "D" RETORT MOUTHPIECE PATTERN (CAST IRON)—TOP RING AND SWIVEL IS WORKED LOOSE—(A) STRICKLE AND CORE BARREL FOR BRANCH CORE.

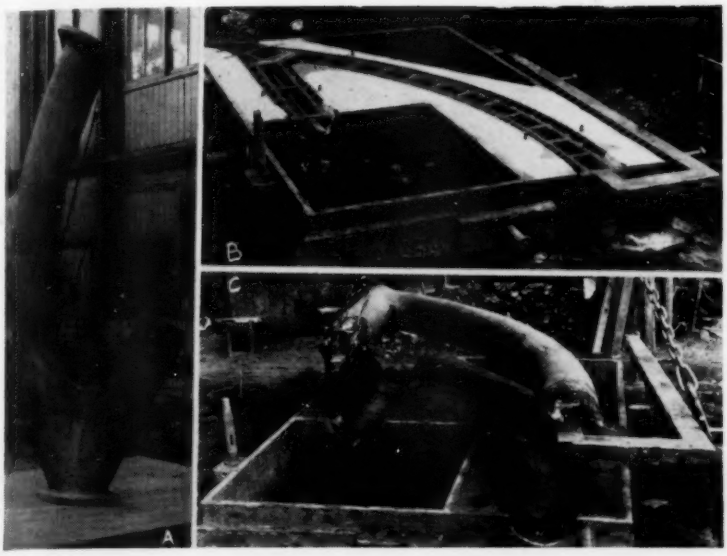

FIG. 4—(A) ARCH PIPE USED ON HORIZONTAL RETORT—MADE WITH GREEN SAND CORE AND FABRICATED CORE IRON, (B) DRAG PART FOR ARCH PIPE, SHOWING SHELL PATTERN AND CORE IRON IN POSITION, (C) GREEN SAND CORE READY TO BE PUT IN POSITION—THE CORE IS 6 IN. IN DIAMETER.

branch. It is guided in and located by pins which screw in the box side.

8. Fig. 4 illustrates an arch pipe also used on this horizontal type of gas retort. From a foundry point of view, it illustrates the use of a shell pattern, a fabricated core iron, and a green sand core. Further applications of this principle with added details will be discussed later. Photographs which illustrate developments offering great possibilities will be discussed also.

CONTINUOUS VERTICAL RETORTS

9. High temperature carbonization is the backbone of the gas industry today. The vertical retort is its essential medium in England. One type, the Woodall-Duckham continuous vertical retort, is made in three sizes. The retort proper is built of silica brick, but both top and bottom are cast iron. They are in the nature of special service castings, as there is heat, abrasion by coke, and corrosion due to acid attack; and the metallurgically involved question of corrosion at elevated temperatures.

10. Fig. 5A shows the bottom ironwork across the minor axis of the retort. Fig. 5B is a view across the major axis.

11. Consideration of the fact that newly made coke falls directly from the retort through the extractor to the hoppers, there to await discharge, will give some indication of the complex conditions this cast iron section must withstand. A complete description of molding methods, of equipment, of sand control, of metallurgical considerations, would extend beyond the possibilities of this paper. Therefore, the author proposes to deal only with isolated examples which may hold interest for the expert as well as that general fraternity of foundrymen to whom this paper is addressed particularly.

MOLDING METHODS

12. Figs. 6A and B illustrate some of the manufacturing methods employed on certain bottom ironwork castings. The illustrations are self-explanatory, and call for no special comment. The principle is to avoid cores as far as possible. All castings are made in green sand. In Fig. 6A, the pattern, which is of cast iron, is lifted with the outside of the mold, and later drawn downwards. Mold quantities are insufficient to warrant elaborate quantity production methods, as demand is recurring rather than continuous. However, much has been done in the way of shaped molding boxes, special turning-over boards, suitable sand, and a helpful foundry layout to improve the efficiency of production.

13. Many of the castings in a retort do lend themselves to more mechanized production methods. A foundry section exists

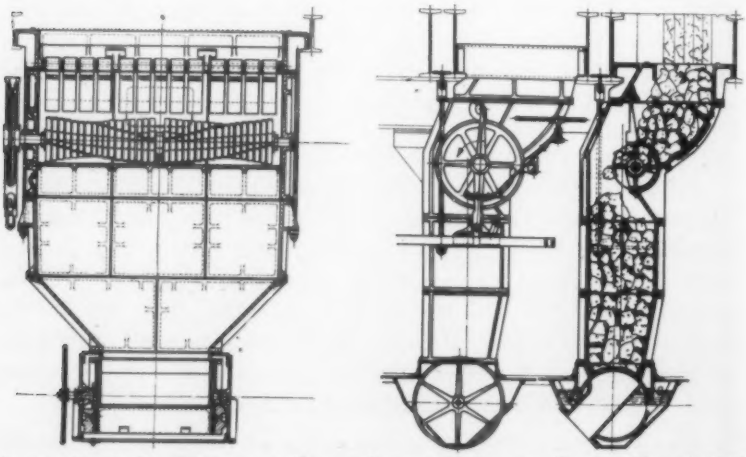

FIG. 5—(LEFT) CROSS SECTION OF BOTTOM SECTION OF VERTICAL RETORT, (RIGHT) CROSS SECTION OF BOTTOM IRON WORKS (MAJOR AXIS).

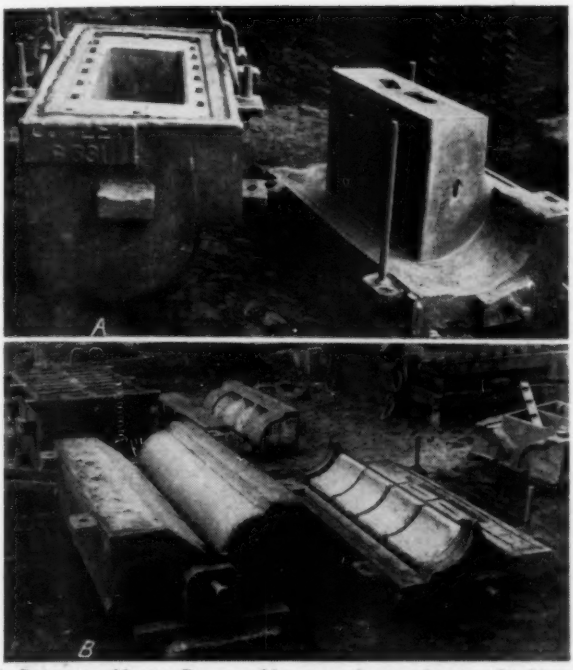

FIG. 6—(A) DRAG AND MIDDLE PART OF MOLD FOR OUTLET SHOOT, (B) MOLD FOR "EXTRACTOR" FRONT CASTINGS.

where cope parts are rammed by a sand slinger, and drag parts by a 10 x 6 ft. jolt ram turn-over machine. In addition, there is a range of work which is rammed entirely by sand slinger and handled on a conveyor system.

14. Fig. 7 gives one example of top ironwork for these continuous vertical retorts. Figs. 8 to 10 are of individual castings, and the production equipment and method which goes with them. Again, both mold and core of all castings are made in green sand. It should be said that individual castings range in weight from 1½ to 12 cwts. (168 to 1344 lb.) and that the molding box shown in Fig. 8 is some 8 ft. long, 4 ft. wide, and when complete, 3 ft. deep.

15. It will be observed that sometimes the structural unit is cast in one piece and sometimes in sections and later assembled. This is because of operating conditions as well as manufacturing considerations. The illustrations cover both types.

16. The making of green sand cores is a practice in which much experiment is yet to be done. The cheapness of the method

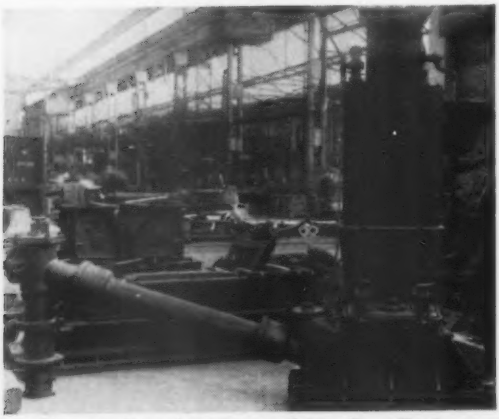

FIG. 7—VIEW OF PART ASSEMBLY OF SMALL TYPE TOP-IRONWORK.

on production of this kind is undeniable and obvious, but the greatest difficulty is to achieve satisfactory results without a long apprenticeship by the molders concerned on this type of work. Nor is it just a question of economics as cooling rates are important. Once a composition has been established, a green mold with a dry core may carry a handicap metallurgically.

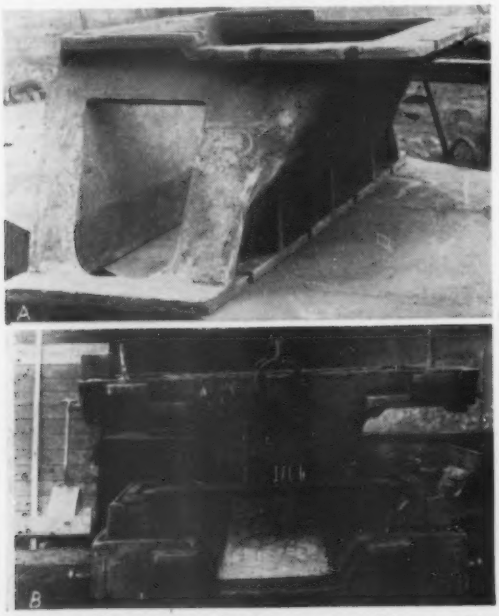

FIG. 8—(A) HOPPER-FOOT CASTINGS FOR TOP-IRONWORK ON VERTICAL RETORT, (B) MOLD FOR HOPPER-FOOT READY TO CLOSE.

17. In the production sphere, the essential need in green sand core practice is a suitably designed, adequately fitted, core iron. Both here and in later illustrations, it will be observed that the core iron sections are prevented from twisting by tie-pieces, and are bound together as a whole by screwed bars.

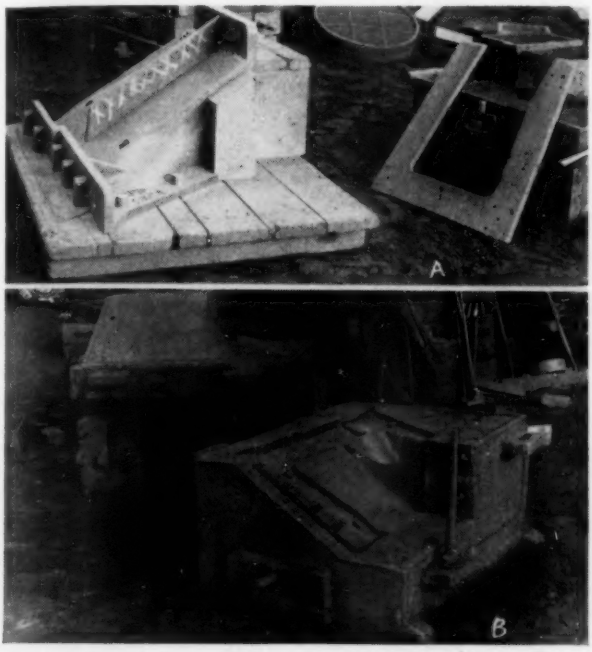

FIG. 9—(A) PATTERN FOR END PLATE OF HOPPER-FOOT BUILT UP IN SECTIONS, (B) MOLD FOR HOPPER-FOOT END PLATE.

18. In our plant, the cores are made alongside the mold whenever possible. After casting, the core-iron is taken out by liberating the various sections and later re-assembled for further use.

COKE OVEN CASTINGS

19. Figs. 11 and 12A and B are of coke oven reversing valves, and illustrate still better the production of gas plant castings in green sand molds and cores. Here special provision had to be made in the core box for the fact that the core is to be made green. Fig. 12B illustrates this clearly. The difference is that the insides of the core box are fitted as loose pieces. Usually no special provision is necessary.

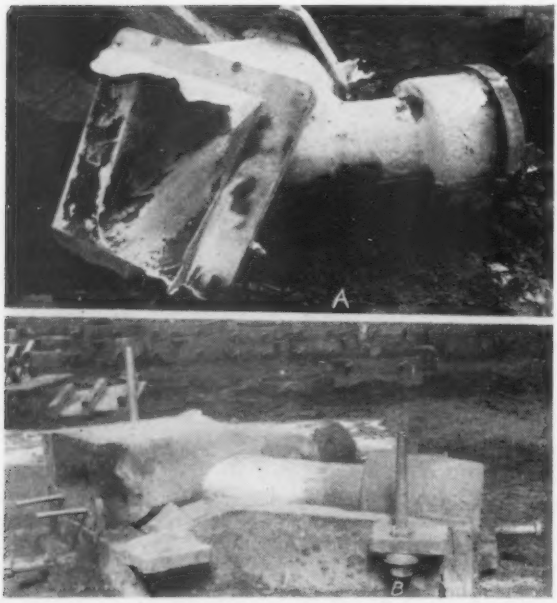

FIG. 10—(A) GAS OFFTAKE CASTING FOR VERTICAL RETORT, (B) GAS OFFTAKE MOLD WITH GREEN SAND CORE IN POSITION.

20. The particular reason can best be seen by reference to the mode of making these cores. The procedure is to line the bottom half of the core box with sand (about ¾ in.), bed in the core iron and ram sand up to the joint of the box. The bottom half is then vented by a thin wire, bars put on to act as a center vent, and the top half of the box placed in position. Sand is rammed in from the open end of the box until the whole is firmly packed. The top half of the box is taken off, the loose sides in the bottom half of the core box drawn out, and the core is then ready to be lifted away. After finishing and sleeking the core is placed in the mold.

FIG. 11—GROUP OF COKE-OVEN REVERSING VALVE CASTINGS.

21. The end extensions on the core-irons serve as lifting bars, and also as a means of holding down the core. No studs or chaplets are necessary. There is a core-iron in the bottom half of the core only.

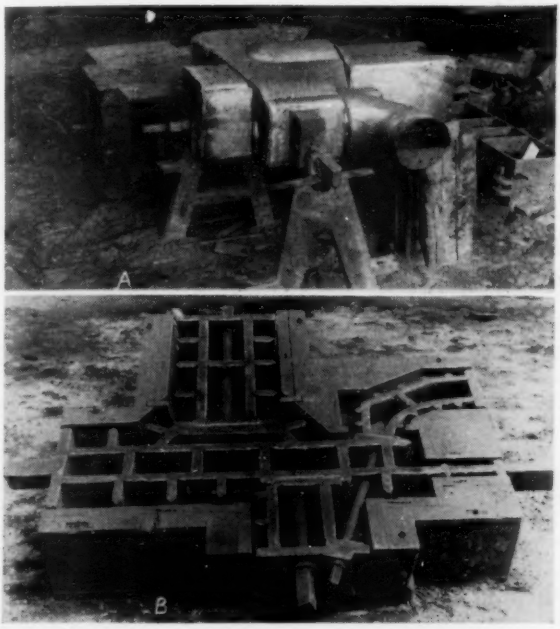

FIG. 12—(A) GREEN SAND CORE FOR REVERSING VALVE BODY CASTING, (B) CORE-BOX AND FABRICATED CORE-IRON FOR REVERSING VALVE BODY CASTING.

22. Other relevant details are that these larger green sand cores are allowed to air dry over night before use whenever possible; usually the molder makes the core as well as the mold. Part of the provision for this type of molding is properly shaped molding boxes into which the core irons can fit and be held in position.

23. The arbor portion of the core iron always carries a larger bar, which fits accurately into a place in the molding box left specially for it. The construction adopted allows a single core iron to be used hundreds or even thousands of times. Sand for the cores is of the same composition as that for molds.

SAND DATA

24. Throughout this foundry, none of the sand for green sand

work is milled. A naturally bonded sand is used. Mixing is effected either by hand or by putting the sand through a pin-type disintegrator. About 10 per cent coal dust of superfine grade (passed through 200 mesh) is used. The ratio of new sand varies from 5 to 30 per cent according to the type of castings.

Table 1

SAND ANALYSIS OF NEW YELLOW SAND

	Dried Sand as Received	*Clay Substance*
Loss on Ignition	2.00	12.71
Silica	80.92	36.54
Ironoxide	1.57	10.28
Alumina & Titania	6.87	31.02
Lime	0.40	1.00
Magnesia	....	....
Alkali oxides	2.76	3.45
Refractoriness	1710°C.	
Sinter	1340°C.	

25. Table 1 and Fig. 13 cover the general properties of new facing and backing sand. This sand is used throughout the green sand system, for hand molding, jolt-ram machine, sand-slinger, and squeeze machine. It leaves a good face on the castings even with rapid production methods. Castings are not sand blasted because of the question of skin removal. This reluctance to sand blast prevents the use of certain types of synthetic sand which do not leave as desirable a skin as could be wished. Molds are dusted with plumbago and sleeked.

METALLURGICAL PROBLEMS

26. Discussion of the metallurgical problems of castings for gas production is hampered by the wide diversity or service conditions encountered. In a single modern gas plant, one could find almost every type of cast iron from the alloyed, wholly austenitic, to the most common phosphoric. Even in a single retort, regard is paid to localized conditions, except where "mutually antagonistic requirements necessitate a compromise."[1]

27. The castings illustrated come under different specifications. The same type of casting may be manufactured to four or

[1]Pearce, J. G., *"Cast Iron, Research and the Gas Industry,"* Institution of Gas Engineers—Publication No. 183-68.

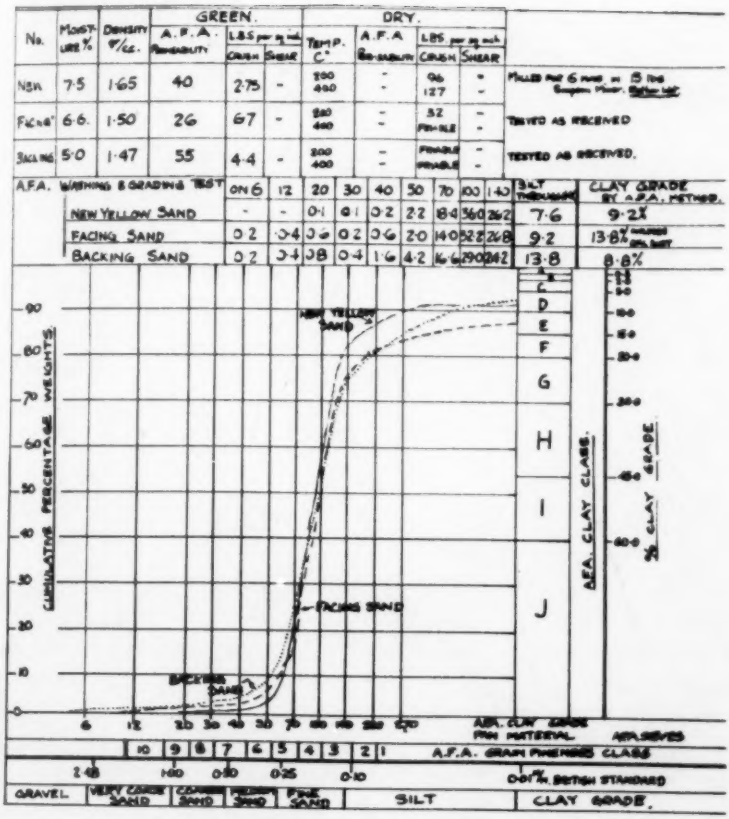

FIG. 13—SAND PROPERTIES DATA SHEET.

FIG. 14—CASTINGS FROM WHICH TEST-BARS (T) WERE TAKEN. ALL TESTS WERE FROM APPROXIMATELY THE SAME POSITION.

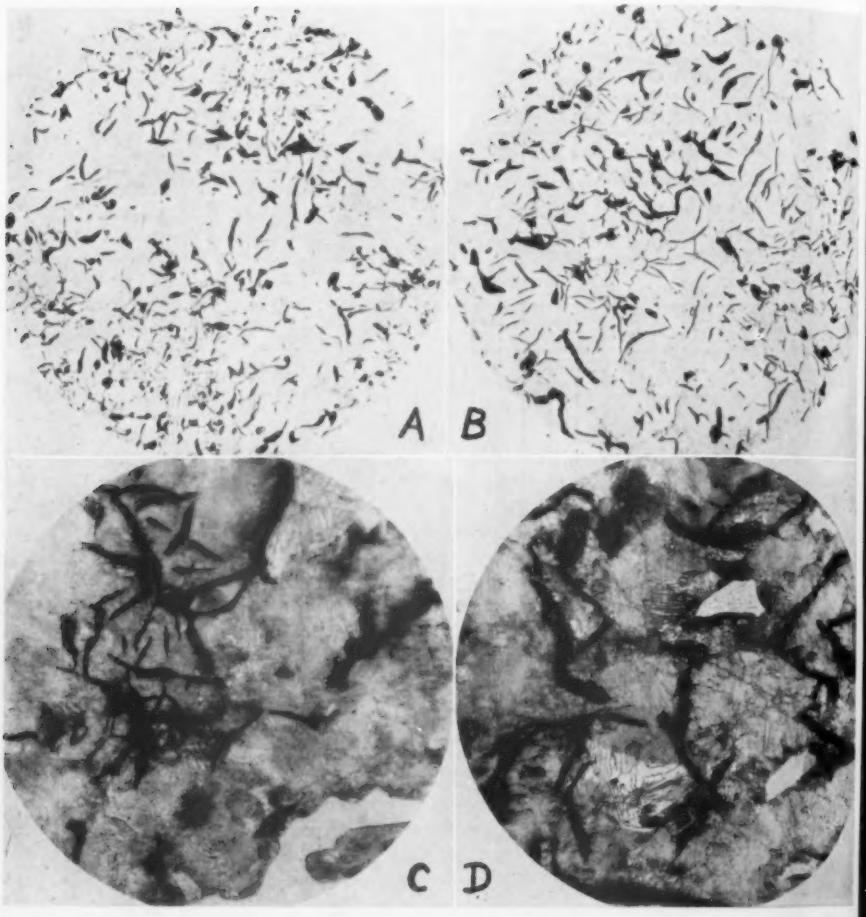

FIG. 15—MICROGRAPHS FROM SECTIONS OF BAR ONE (0.875 IN. DIAMETER) BAR CAST IN DRY SAND MOLD. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

five different specifications, all of which have reference to local conditions encountered. These conditions may be in Cornwall and not in Scotland, or in Wales and not in Ireland. These conditions often concern the sulphur content of the coal as well as the type of water available locally, as in the process of gas making, steam is introduced at certain points in the bottom ironwork, and the reactions are often complex and involved.

28. It follows that foundry control, which is becoming increasingly common, has to be directed not only to composition and

physical properties, but also to structure as affected by mold materials and cooling conditions.

29. Recent experience in the handling of what might be regarded as "sensitive" alloys, sensitive, that is, to mold conditions and temperature changes, indicate that test bar results should be checked periodically where possible against the castings manufactured for sale. Particular regard is paid to the conditions of test-bars and their manufacture if they are to serve as more than

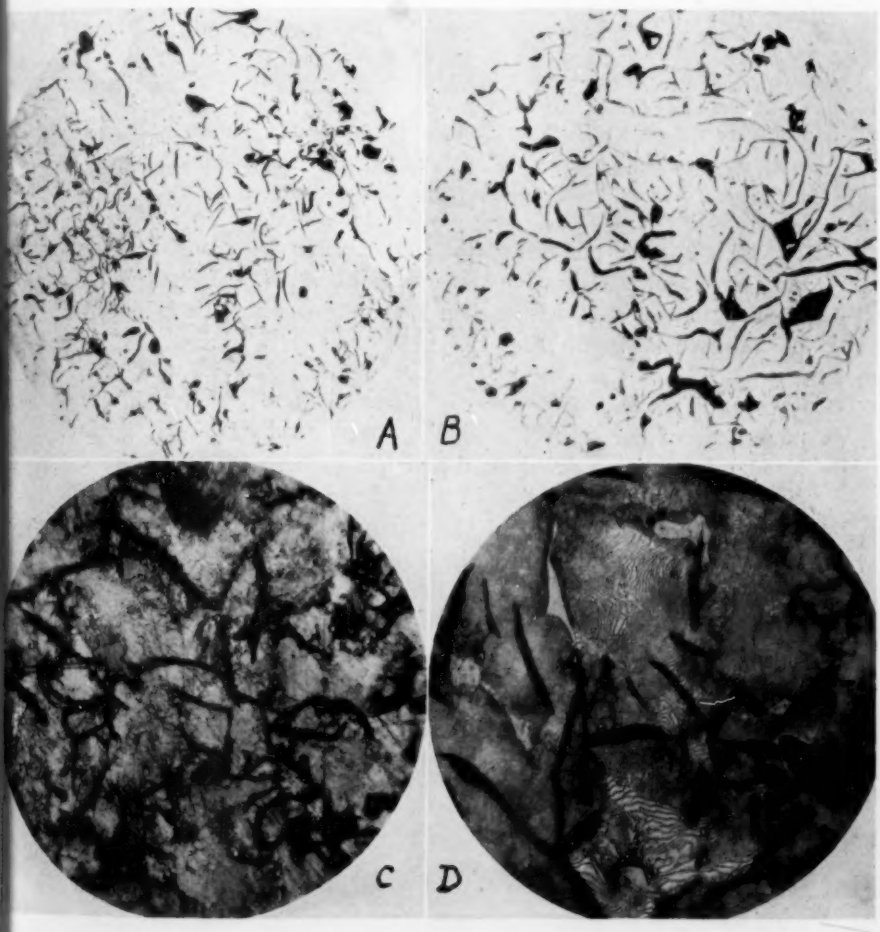

FIG. 16—MICROGRAPHS FROM SECTIONS OF BAR TWO (0.875 IN. DIAMETER). BAR CAST IN GREEN SAND MOLD AND FROM SAME LADLE OF METAL AS FIG. 15. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

Table 2

BRINELL HARDNESS TESTS

Test Piece		*Edge*	*Center*	
Inch	*Bar No.*			
0.875	(1)	269	228	5 m.m. Ball
0.875	(2)	286	228	750 k.g. Load
0.875	(3)	332	241	30 secs.
1.2	(1)	248	217	
1.2	(2)	248	228	
1.2	(3)	286	248	10 m.m. Ball
Casting	*Top*	*Bottom*		3.000 k.g. Load
No. 1	218	220	200	30 secs.
No. 2	232	232	207	
No. 3	253	251	221	

nominal indications of the quality of material. Tables 2, 3 and 4 indicate results of a test of this kind, and the photomicrographs of Figs. 15 to 23 further elucidate the position.

30. For the purpose of this test, six test bars and three castings were poured from the same ladle of metal. Afterwards, test pieces and test bars were taken from the castings themselves, in the manner shown in Fig. 14.

31. The test bars were cast vertically and individually, and poured from the top. The castings were numbered 1, 2 and 3. No. 1 was cast in a dried mold and left in the sand 12 hrs. No. 2 was cast in a green sand mold and left in the sand 12 hrs. No. 3 was cast in green sand and the top part lifted off 10 min. after the casting was poured. A Brinell reading was taken on the pieces sectioned from the castings on the top face, the bottom face, and the center. The section was ¾ in. The castings were poured in the

Table 3

TRANSVERSE TESTS

Test Piece		*Distance between Supports,*	*Deflection,*	*Diameter at Fracture,*	*Breaking Load*	*Transverse Rupture Stress*
Inches	*Bar No.*	*Inches*	*Inches*	*Inches*	*Lb.*	*tons per sq. in.*
0.875	(1)	12	0.14	0.845	1437	29.50
0.875	(2)	12	0.13	0.879	1418	29.00
0.875	(3)	12	0.12	0.868	1358	27.70
1.2	(1)	18	0.21	1.180	2330	27.70
1.2	(2)	18	0.20	1.130	2283	27.00
1.2	(3)	18	0.22	1.180	2547	30.26

Table 4

TENSILE TESTS

Test Piece		*Diameter at Fracture,*	*Load,*	*Load, Lb.*
Dia. In.	*Bar No.*	*Inches*	*Lb.*	*per. sq. in.*
0.875	(1)	0.563	8,380	33,700
0.875	(2)	0.559	8,300	36,100
0.875	(3)	0.564	9,690	38,700
1.2	(1)	0.794	15,600	31,500
1.2	(2)	0.794	15,680	31,700
1.2	(3)	0.794	16,760	33,900
Bar from Casting	(1)	0.554	7,940	32,900
Bar from Casting	(2)	0.552	8,850	37,100
Bar from Casting	(3)	0.562	9,480	38,500

order 1, 2, 3. Temperature at cupola spout was approximately 1400°C. (2552°F.). The pouring temperature was approximately 1350°C. (2462°F.).

32. Two sizes of test bar were used, the British Standard 1.2 in. and the 0.875 in. Three of each made up the six and were marked 1, 2 and 3. The pair of bars marked 1, i.e. one large and one small, were cast in dried molds and left in sand 12 hrs. No. 2 pair of bars were cast in a green mold and left in the sand 12 hrs. No. 3 pair of bars were cast in a green sand mold and taken out when they had reached a temperature of approximately 950°C. (1742°F.) and cooled rapidly by an air blast. Test bars marked 1, 2 and 3 were cut from the castings 1, 2 and 3. The physical test and the microstructures (Figs. 15 to 23) are from these test pieces.

33. Transverse tests were taken in the "as-cast" condition, and the only cleaning treatment was to brush off adhering sand. Tensile tests were taken on one half of the broken transverse test bar.

34. It will be seen that neither 1.2 in. or 0.875 in. size test bar gives a complete picture of the conditions of the casting. However, the result of the tensile strength of the 0.875 in. bar approximate closely the results obtained from bars taken from the castings themselves. These bars were machined to the same size as the test bar. More particularly it will be observed that differences in treatment

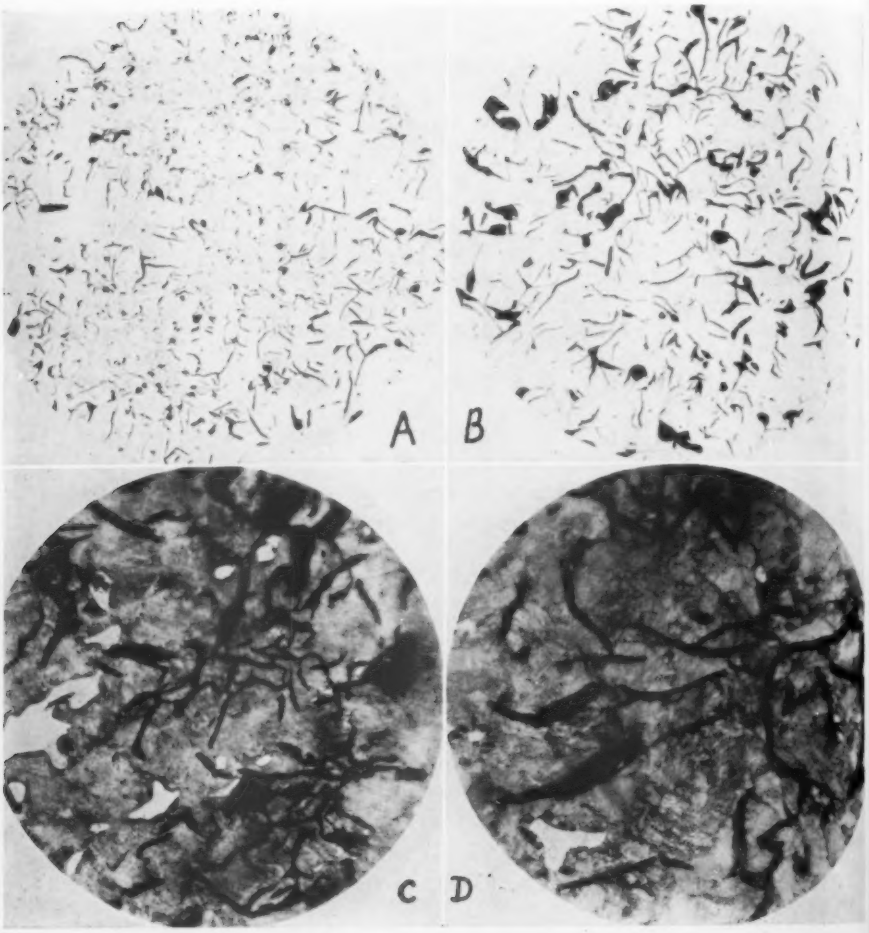

FIG. 17—MICROGRAPHS FROM SECTIONS OF BAR THREE (0.875 IN. DIAMETER) BAR CAST IN GREEN SAND MOLD AND COOLED BY AIR BLAST FROM 950°C. (1742°F.) METAL FROM SAME LADLE AS FIG. 15 AND 16. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

are reflected alike in both test bar and casting to roughly the same degree. Though it cannot be said that the Brinell tests show any close relationship between casting and test bar. No transverse tests were taken from the sectioned castings because it was felt that machined bars would be hardly likely to give a true reading. It was not intended that future practice should cater for the machining of all transverse test bars.

35. The real point at issue was to what extent test bars could be held to give an adequate picture of the castings for sale, and in what degree changes in treatment could be reflected in test bars, assuming the composition of the metal to remain constant.

36. The composition of the metal for the foregoing tests was: total carbon 3.21 per cent, silicon 1.28 per cent, manganese 0.89 per cent, phosphorus 0.32 per cent, sulphur 0.074 per cent.

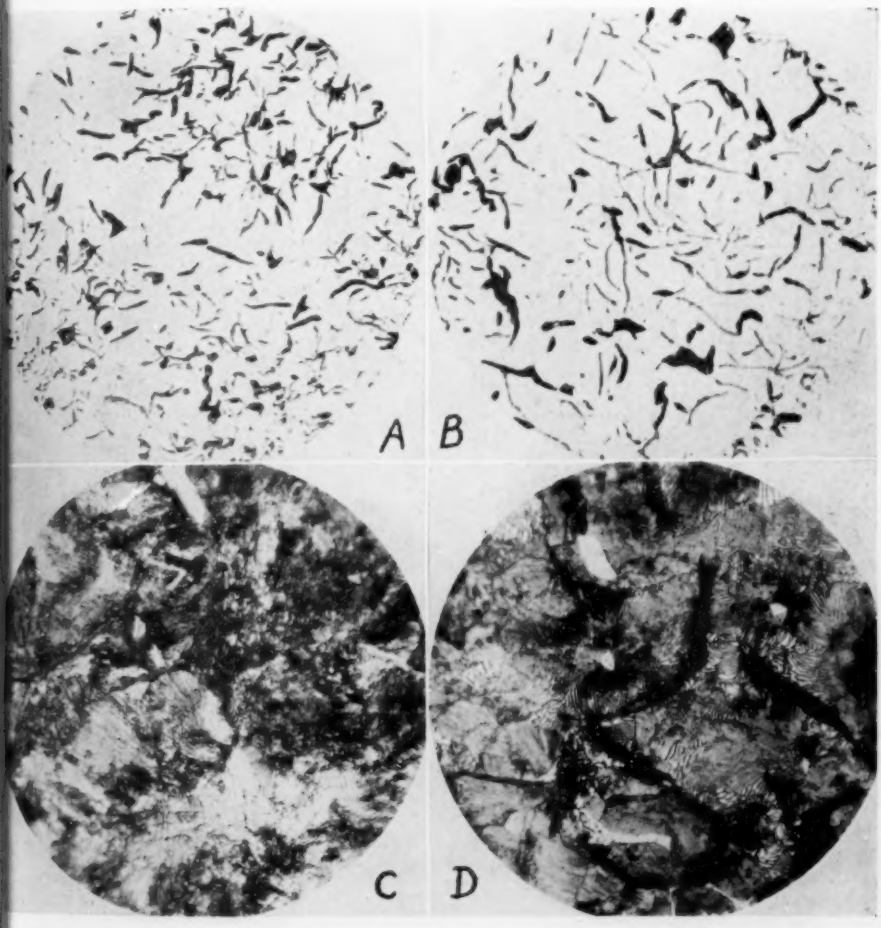

FIG. 18—MICROGRAPHS FROM SECTIONS OF BAR ONE (1.2 IN. DIAMETER.) BAR CAST IN DRY SAND MOLD AND FROM SAME LADLE OF METAL AS FIG. 15, 16 AND 17. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

Similar experiments are in hand on other types of metal, and an investigation is also being made into the effect of moderate temperatures on the modifications of structure induced by mold materials and foundry conditions.

37. It is not held that analysis has lost its place, nor controlled melting, but mold materials and mold conditions, particularly in green sand practice with sensitive irons, are factors to which

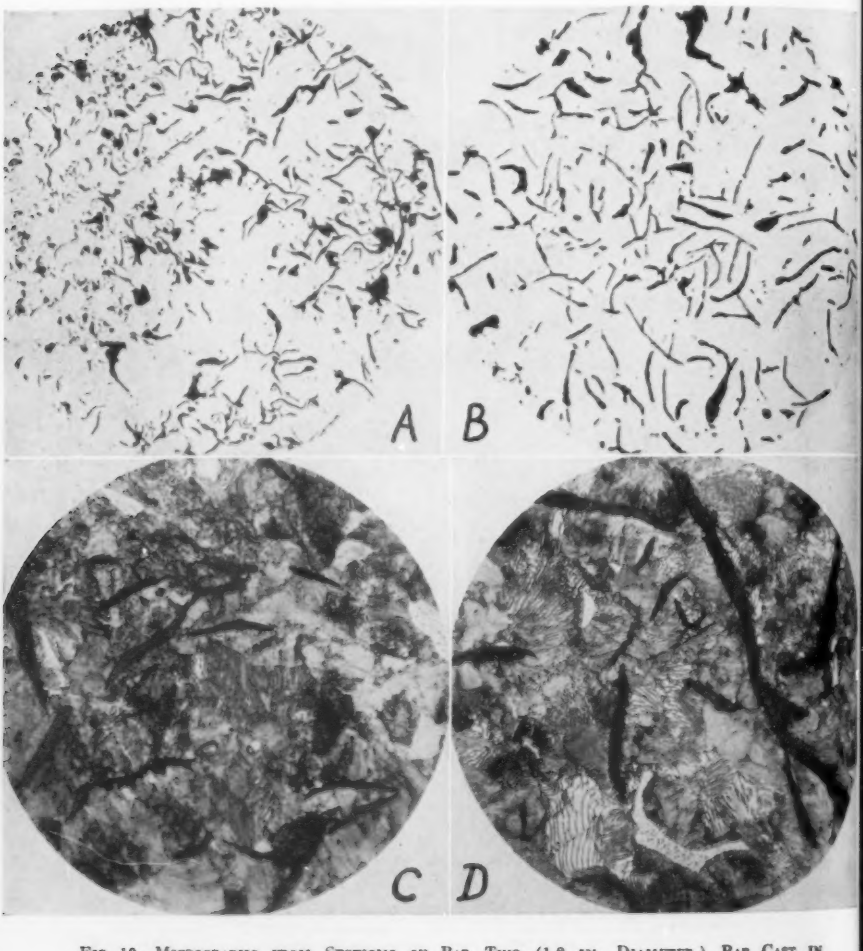

FIG. 19—MICROGRAPHS FROM SECTIONS OF BAR TWO (1.2 IN. DIAMETER.) BAR CAST IN GREEN SAND MOLD AND FROM SAME LADLE OF METAL AS FIG. 15, 16 AND 17. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

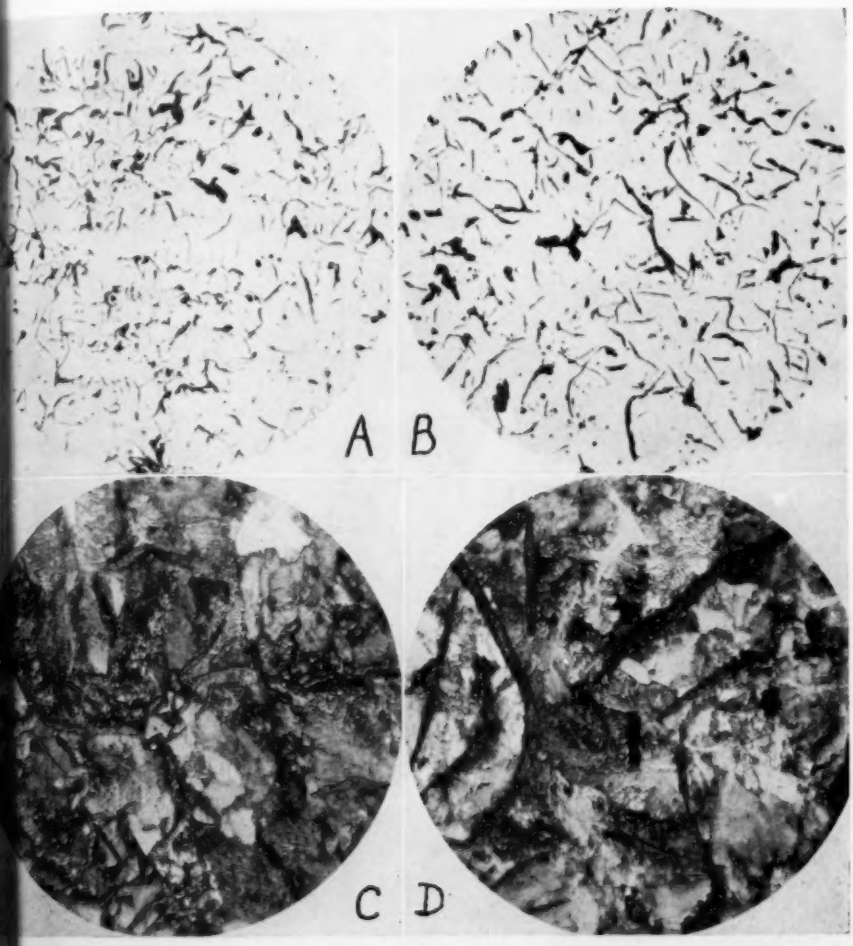

FIG. 20—MICROGRAPHS FROM SECTIONS OF BAR THREE (1.2 IN. DIAMETER.) BAR CAST IN GREEN SAND MOLD AND COOLED BY AIR BLAST FROM 950°C. (1742°F.). METAL FROM SAME LADLE AS FIGS. 15, 16 AND 17. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 100. (D) CENTER, ETCHED, X 100.

scarcely enough attention has been paid. This applies with particular force where, as in the manufacture of gas plant castings, not only engineering properties must be obtained, but also a degree of heat resistance and corrosion resistance which strength figures do nothing to indicate. The micro-structures referred to, form an interesting commentary on this problem.

LOW TEMPERATURE CARBONIZATION

38. A sphere in which cast iron seems to be taking its place as a retort material is that of low temperature carbonization. There are many examples of these in England.

39. Fig. 24A illustrates a pattern of a type of casting used for building these retorts. They do, of course, operate vertically. It will be seen that, in the illustration given, an oval section retort

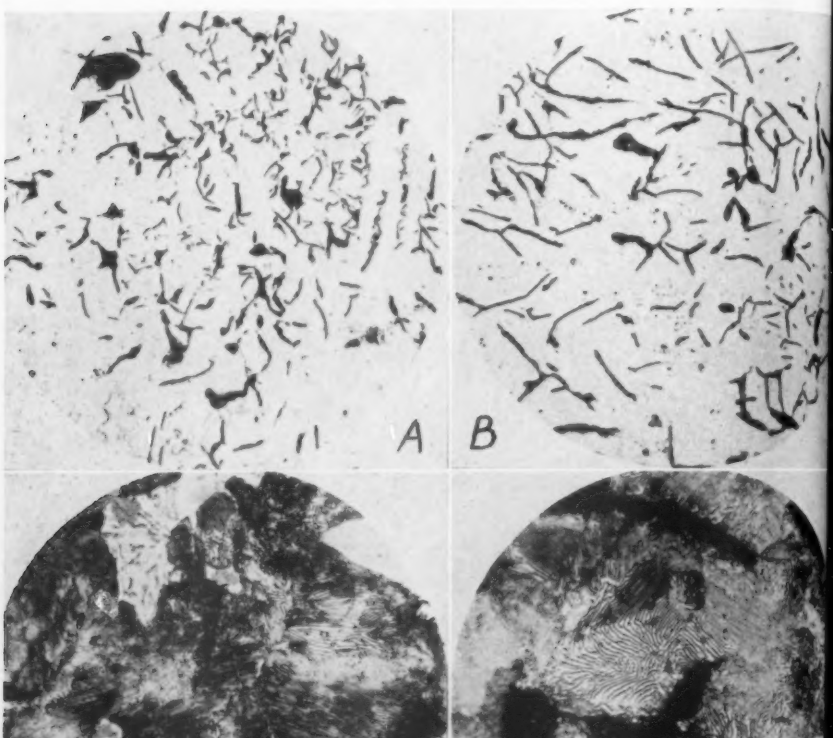

FIG. 21—MICROGRAPHS FROM SECTIONS OF TEST PIECES TAKEN FROM CASTING POURED FROM SAME LADLE OF METAL AS BARS OF FIG. 15, 16 AND 17. POSITION OF TEST PIECE SHOWN AS T OF FIG. 14. TEST PIECE ONE CAST IN A DRY SAND MOLD. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

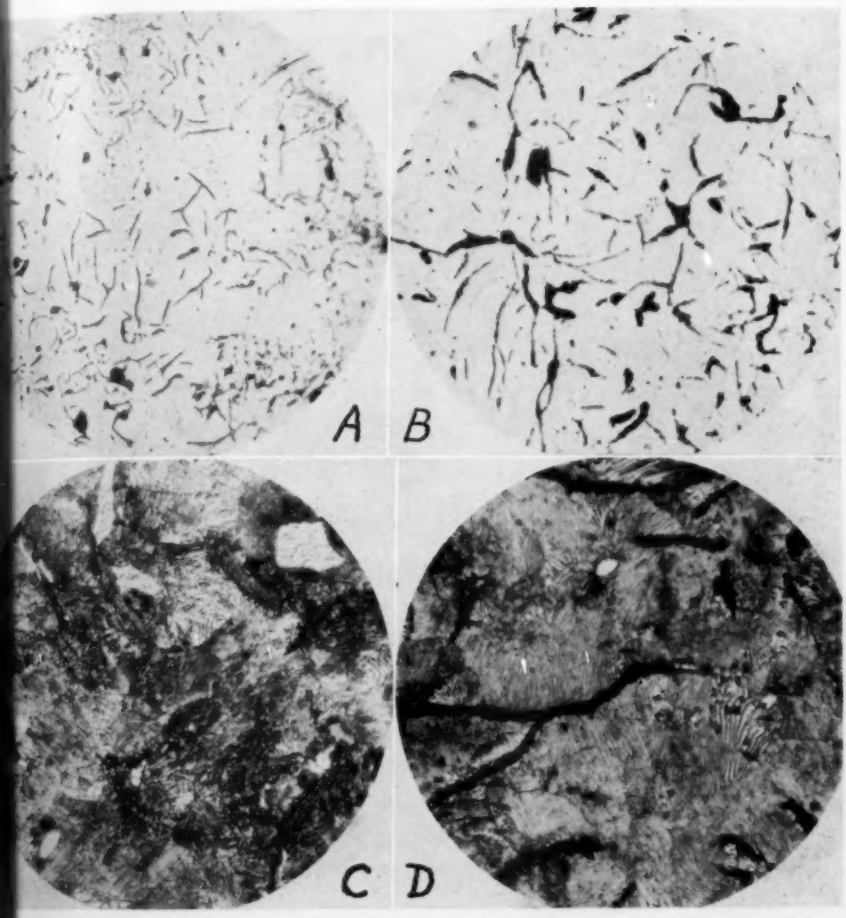

FIG. 22—MICROGRAPHS FROM SECTIONS OF TEST PIECES TAKEN FROM CASTING POURED FROM SAME LADLE OF METAL AS BARS OF FIG. 15, 16 AND 17. POSITION OF TEST PIECE SHOWN AS T OF FIG. 14. TEST PIECE TWO CAST IN A GREEN SAND MOLD. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

can be formed by bolting together the separate sections. The operating conditions are such that heat is applied on the outside and serves to coke the coal on the inside.

40. The operating temperature for this process is approximately 650 to 700°C. (1202 to 1292°F.). It is imperative that the plates which form the retort should not only have long life, but should neither distort nor crack. Given these conditions, cast iron for low temperature carbonization has definite advantages over all other

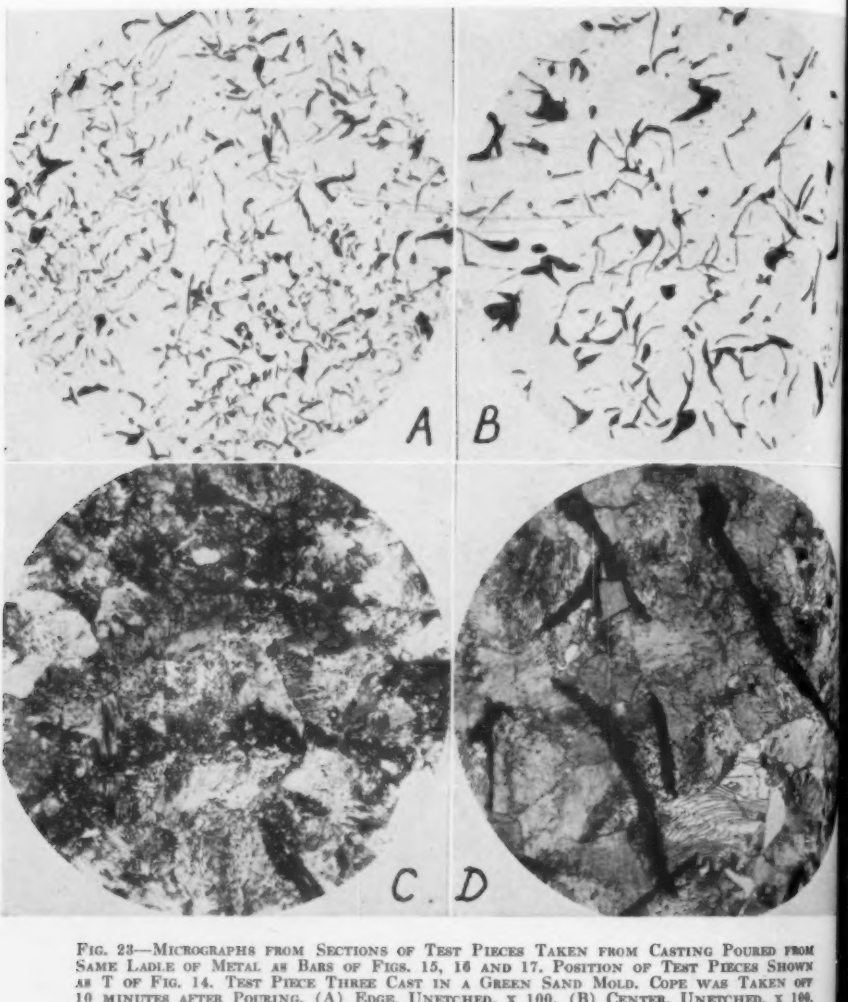

FIG. 23—MICROGRAPHS FROM SECTIONS OF TEST PIECES TAKEN FROM CASTING POURED FROM SAME LADLE OF METAL AS BARS OF FIGS. 15, 16 AND 17. POSITION OF TEST PIECES SHOWN AS T OF FIG. 14. TEST PIECE THREE CAST IN A GREEN SAND MOLD. COPE WAS TAKEN OFF 10 MINUTES AFTER POURING. (A) EDGE, UNETCHED, X 100. (B) CENTER, UNETCHED, X 100. (C) EDGE, ETCHED, X 500. (D) CENTER, ETCHED, X 500.

available materials at competitive prices. It is a field in which much can be accomplished.

41. Production in this case is from dried molds, and given selected raw materials and controlled melting, the foundry control rather follows the chilled roll foundry practice, inasmuch as no casting is poured until a satisfactory chill test has been obtained. Distortion in the foundry is dealt with by special handling. Also a stress relief treatment at 650°C. (1202°F.) is necessary previous

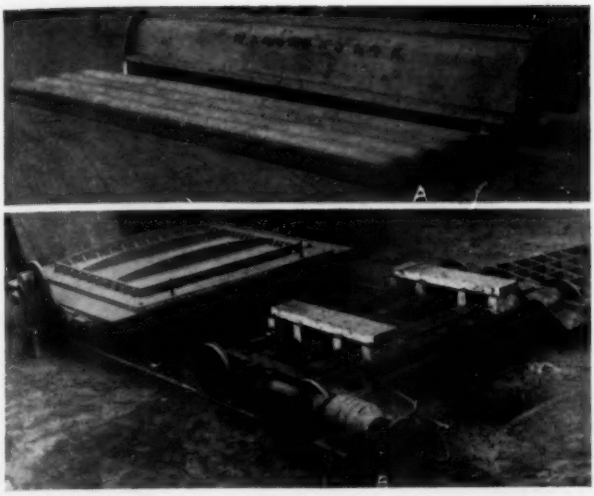

FIG. 24—(A) PATTERNS FOR TYPE OF LOW TEMPERATURE—CARBONIZATION RETORT—CAST WITH SPECIAL HEAT-RESISTING IRON. (B) JOLT-RAM ROLL-OVER MACHINE FOR RAMMING DRAG-PART MOLD FOR PURIFIER SECTIONS.

to machining to insure satisfactory working service. Microphotographs Nos. 25A and B show two stages in the life of these castings. The analysis of the casting is: Total carbon 3.17 per cent, silicon 1.1 per cent, manganese 0.72 per cent, phosphorus 0.10 per cent, sulphur 0.089 per cent, chromium 0.5 per cent. The tensile test on a bar taken from the casting previous to machining was 16 tons per sq. in. (35,840 lb. per sq. in.).

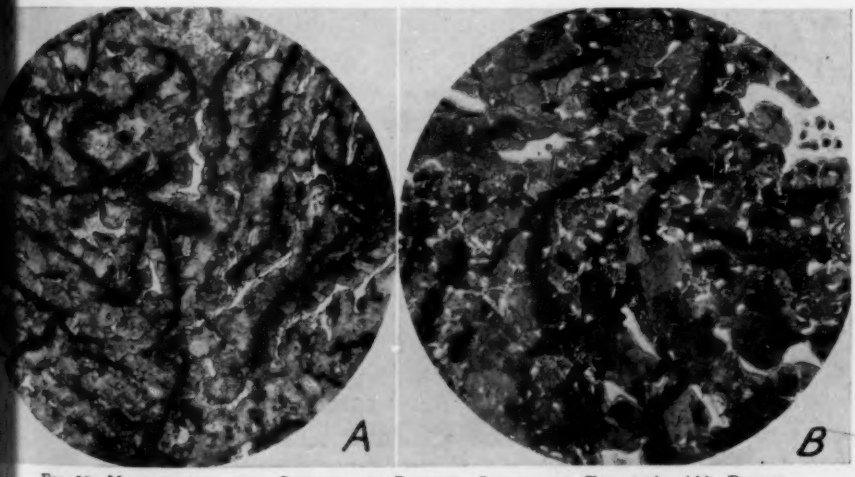

FIG. 25—MICROGRAPHS FROM CASTINGS OF PATTERN SHOWN IN FIG. 24A. (A) BEFORE GOING INTO SERVICE—ETCHED, X 100. (B) AFTER 2½ YEARS IN SERVICE—ETCHED, X 100.

FIG. 26—GAS PURIFIER BOXES OF CAST IRON AT BIRMINGHAM.

42. There are, of course, castings in other systems of low temperature carbonization not illustrated here. The problem there, as here, is not only one of composition, but design and operating as well as manufacturing conditions.

GAS PURIFICATION

43. Gas purifier boxes are still a job for the foundry in England. The special corrosion resisting properties of cast-iron, together with the improved physical properties now obtainable, insure a continuance of demand for at least some years to come. Fig. 26 shows part of a typical installation at Birmingham and Fig. 24A shows the jolt-ram roll-over machine for the production of purifier box sections.

44. The plates forming the bottom of these boxes, which are 40 ft. x 40 ft., are 5 ft. square and the side plates are 6 ft. 6 in.

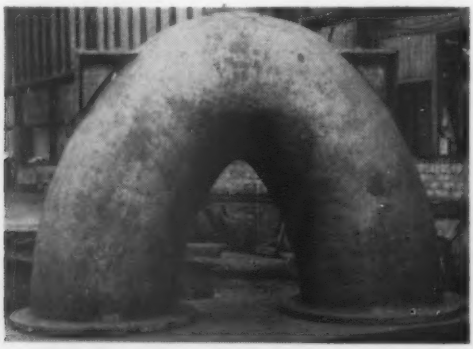

FIG. 27—HYDRAULIC VALVE FOR GAS PURIFIER BOXES MADE IN GREEN SAND—WEIGHT APPROXIMATELY 2800 POUNDS.

x 5 ft. Production is completely mechanized. The sand used is that of which particulars have been given earlier. All molds are cast in green sand.

45. Fig. 27 is of the hydraulic valve which can just be seen in the general photograph of gas purifier boxes given in Fig. 26. This again is an illustration of the use of a green sand core and a fabricated core-iron. Fig. 28A and B will make the equipment

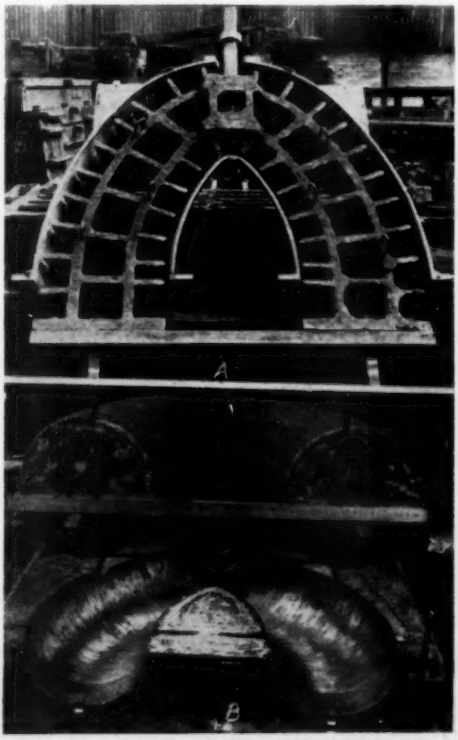

FIG. 28—(A) CAST IRON SHELL PATTERN AND FABRICATED CORE-IRON FOR MAKING HYDRAULIC VALVE. (B) MOLD AND GREEN SAND CORE FOR HYDRAULIC VALVE READY FOR CLOSING.

provision for this kind of production quite clear. The diameter of the valve is 24 in. and the weight approximately 2,800 lb.

46. Fig. 29 shows elements of a special type gas-heater fixed in the pipeline to the purifier boxes and used to control the temperatures of gas for its more rapid purification. The needle heater elements, which form the medium for heat transference, are of cast-iron and have needles in the bore of the pipe as well as those which are obvious on the outside. These castings are made on a

FIG. 29—SPECIAL TYPE GAS HEATER ELEMENTS READY FOR INSERTION IN GAS HEATER BODY.

jolt-strip machine, and so far as the mold itself is concerned are cast in green sand. The inside is formed by oil sand cores.

47. All castings for gas purification are cast in a metal giving a tensile test on a 1.2 in. bar of 15 to 16 tons per sq. in. (33,600 to 35,840 lb. per sq. in.). A normal micro-structure is approximately the one shown in Fig. 20.

CONCLUSION

48. It will be realized that this survey of castings for the British gas industry does little more than touch the fringe of the matter. Much could be said about the intermittent vertical retort, the modern horizontal retort and other types of continuous vertical retorts than that used for illustration. If the nature of the foundry work for this industry has been made clear, or if the problems indicated stimulate interest the author is satisfied. Though he does not forget that he leaves the vast field of castings for gas utilization plant to other hands.

49. Finally, thanks are due to The Woodall-Duckham Company, London, and to Newton, Chambers & Company, Sheffield, England, for permission to prepare this paper and to R. C. Tucker, M. A., chief metallurgist, Newton, Chambers & Company for his personal interest and help.

DISCUSSION

Presiding: FRED J. WALLS, International Nickel Co., Detroit

CHAIRMAN WALLS: The next paper is the official exchange paper from Great Britain. We are indeed sorry that the author, Mr. Whitehouse, is not here to present the paper to us, however, we are very fortunate in having a man from across the water who has distinguished himself in this industry. On behalf of Mr. Whitehouse, he is to present

this paper which is entitled "Castings for the British Gas Industry." The paper will be presented by our good friend, and a past President of the Institute of British Foundrymen, J. E. Hurst, Director of Research, Bradley & Fostêr, Darlaston, England.

MR. HURST: While Mr. Whitehouse could not come to this conference, it so happened that I was to be here and the Council of the British Institute asked me if I would do them the honor of presenting the paper to you. Now, the firm with which Mr. Whitehouse is connected is one of the very old British iron works. In fact, it was established in its present position in the year 1793 and has continually operated on its present site since that date and the ownership has remained in the same family. Actually, the concern was established some years before 1793. So it can be regarded as one of the very old iron works. This firm was early engaged in the manufacture of materials for the gas industry—and it is correct to say that it can be regarded as the pioneering firm of the gas light, and possibly the coke industry. It still continues to make castings for the gas industry, the ordinary gas lighting industry and also the coke oven industry.

Mr. Hurst abstracted the paper

CHAIRMAN WALLS: I am sure the American Foundrymen's Association is very grateful not only for the paper, but for the splendid presentation which our good friend from across the water has just made. We always feel honored in having with us past presidents of the Institute of British Foundrymen.

H. B. SWAN:[1] It would be interesting to know further about those heat exchange boxes mentioned as having a chromium content of 30 per cent. What is remaining composition and are they machinable?

H. W. DIETERT:[2] We would like to know the type of mold facing or spray used to protect the sand, that is what is used for facing mold and whether they are skin dried.

MR. HURST: Mr. Whitehouse is the apostle of green sand core making in England and he has taken upon himself quite a missionary job in going around to our various brances of the Institute of British Foundrymen giving papers on the subject of green sand molding. I am sure, Mr. Dietert, that in his written closure the author will give you the data that you want.

MEMBER: What particular claims does the author make for that process or system of molding and core making?

CHAIRMAN WALLS: Mr. Hurst, will you convey to Mr. Whitehouse our appreciation for this very wonderful paper?

F. WHITEHOUSE (*written closure*): In reply to Mr. Swan I would say the composition of what he calls heat exchange boxes and what we call recuperator castings is approximately: Total carbon, 1.8 per cent;

[1] Foundry Superintendent, Cadillac Motor Car Co., Detroit, Mich.
[2] President, H. W. Dietert Co., Detroit, Mich.

silicon, 0.5 per cent; manganese, 0.4 per cent, phosphorus, 0.1 per cent max.; sulphur, 0.1 per cent maximum, chromium, 27 to 30 per cent. The flanges are machined, but special tungsten carbide tools are necessary and the speed is low. The flanges also are drilled, but here special inserts are cast in the flanges to obviate difficulties.

Concerning Mr. Dietert's questions: Our facing for green-sand work is finely ground gas-carbon or Ceylon graphite, and in each case it is dusted on the mold by shaking a bag containing this material over the mold. The powder is then pressed down to the mold by sleeking with tools or rubbing with the hand. Sometimes this facing is applied by taking a handful and rubbing it on the mold. The molds are not skin-dried. A spray is not used on green-sand. I would say further that in the case of the heat exchange boxes mentioned above, the molds are dried, and in this case they are sprayed with raw linseed oil before going into the stove, and then after drying are sprayed with silica flour mixed with water and then put back in the stove for drying off.

In reply to Member: The claim made for green-sand practice in both mold and core is that it is a cheaper method of manufacture whenever it is possible to adopt it safely. It is cheaper chiefly because both directly and indirectly the drying of molds and cores costs money. On some jobs the saving is more apparent than on others, but in almost all cases it exists, and it is the author's firm opinion that so far as English practice is concerned, it is worth while inquiring how far the technique of green-sand molding and core-making can be developed. The skill of the molder does, of course, count for something, but a good deal can be done by sand control and a carefully thought out molding technique. That the effort is worth while can be judged by the fact that some of the castings illustrated are 30 per cent cheaper to manufacture in green-sand than in dry. The question of the effect of green-sand on the physical and general service properties of the castings is also worth consideration, as in the writer's opinion mold materials is a subject on which further investigation is definitely due, if not overdue.

Finally, I would express my regret to those present at the reading of the paper that I was not there to answer questions, but if I can, even now, clear up any points or elaborate any aspect of the problem, I shall be glad to do so.

A Scientific Approach to a Foundry Problem

By L. A. Danse,* Detroit, Mich.

Abstract

The author discusses chronic cracking of castings as encountered from 1912 to 1939. Design, planning and control of this trouble is discussed in general while methods used to determine the causes are described in detail. The results of placing thermocouples in various positions in the core and mold are detailed. Interpretations of the various cooling curves are also included.

Introduction

1. The discussion herein related will bear upon a prolific source of trouble with cracked castings, termed, for want of a better designation, as "heat unbalance." This trouble is entirely separate and distinct from the common causes of casting breakage, such as rough handling, or other operating factors. Cracking resulting from hammering hot castings or too severe shakeout, dropping or throwing castings, such as from conveyor to truck, into cleaning machines or mills; from too severe milling, from too tight clamping, or too high pressure in a water test fixture; or from other purely mechanical means or causes should be distinguished from the type of cracking stated, the causes of which will be traced.

2. Even though cracking from mechanical causes may at first appear puzzling, it easily is traceable to its source by close observation. The cracking, which is really a producer of frayed nerves, is the kind whose cause evades apprehension, even after the closest and most exhaustive tracing and scrutiny.

3. This cause may be termed "heat unbalance." It is the problem child of complex design on the one hand and conventional foundry practice on the other. Elimination of this cause calls not only for observation and scrutiny, but close study of the structural details of the design, metal flow, solidification, and cooling characteristics of the related masses of the casting, both in the mold and after shakeout.

* Chief Metallurgist, Cadillac Motor Car Division, General Motors Corp.

Note: Presented before Gray Iron Session of the 43rd Annual Convention, Cincinnati, Ohio, May 16, 1939.

FOREWORD

4. The first experience with chronic cracking of castings to come to the writer's attention was in 1912-13. During this winter, some large cylinder heads (cylinders with 21 in. bore and 30 in. stroke) weighing about a ton apiece gave considerable trouble by coming out of the molds in several pieces. This state of affairs grew steadily worse as the winter progressed until it was necessary to pour three molds to get one casting.

5. The management became not only upset, but frantic as no remedy appeared. Several consulting foundrymen could offer little help, but only criticized the design.

6. In desperation, the problem was called to the attention of the late John Keller, Purdue University, with the hope that his penchant for putting theories to practical use, might help, even though he knew nothing about foundry work. This hunch on the part of the general superintendent was a good one for Mr. Keller went after the job like a hound dog after a rabbit.

7. He looked over the broken castings, studied molding and pouring conditions and asked a few questions about cast iron and its melting and pouring. He then told some of the junior supervision to get some pyrometers, place them in the next mold, and take records of the pouring and cooling conditions. This was done, the casting poured, and the record made. The casting, incidentally, cracked as usual.

8. Mr. Keller carefully plotted the chart curves on straight line paper (in those days we had nothing else but round smoked chart recorders). From the results, he pointed out that the rates of cooling of adjacent sections of the casting in the mold were such that some sections were shrinking while others were expanding. The logical result of such conditions was just what we had—cracked castings.

9. Thus, Mr. Keller's common sense gave the answer after those of us who were more familiar with the job had overlooked it.

10. Better still, he suggested that the cooling rate could be controlled by proper arrangements of cooling coils in some portions of the molds and heating apparatus in other parts so that all por-

tions of the casting would cool at uniform rates. When this was tried, it showed definite improvement. When it was worked out, it resulted in a complete cure and entire freedom from the trouble.

11. Mr. Keller's example was never forgotten. It was a splendid evidence of the old truth that before a trouble can be fixed, it must first be disclosed, and that after it is found, the cure almost suggests itself.

A New Design

12. The years passed, until a new design of V8 cylinder-crankcase block casting went into production in mid-summer of 1935. Although a complex casting, it started reasonably well and was attended by no un-toward circumstances. In fact, no unusual trouble attended the making of the first several hundred castings.

13. In October, the weather suddenly turned frosty and cold with the result that two-thirds of the castings were cracking in the mid-section and production schedules were disrupted completely. The foundry was not able to produce castings fast enough to keep the machine shop line going.

14. It took only a short time to bring a grin of recollection to the face of the material trouble-shooter. Remembering his experience of 20-odd years ago when, as a youngster, he had been privileged to help Mr. Keller on the big cylinder head job, he dug into the problem.

15. What then took place can be summarized from the report of the material man:

> "High scrappage, from cracks running down into web of center bulkhead from cored camshaft bearing hole, was due directly to an abrupt change in the weather. The temperature dropped about 30°F. and a strong north wind aggravated the condition, resulting in a considerably accelerated cooling rate and consequent high stresses. Cracking naturally followed.
>
> "From the appearance of the cracks, it was concluded at once that cooling stresses were the cause. The remedy was to retard cooling and the quickest way was to cast the camshaft bearing solid, not cored. This gave immediate relief and stopped the scrap. However, boring out the bearing hole from the solid metal was a slow and expensive operation. Risers, or breakoff bobs, were placed on the sides of the bearing boss

and the cores were shaped out by hand until the coremaking equipment could be altered. This also gave good results.

"However, cracks in the bottom water jacket wall, adjacent to the center bulkhead, indicated that changes should be made at these points. Two more ribs were added to the one already in each of the long flat walls. This improved conditions somewhat.

"Cracks in the center bearing support were helped by adding a tie-bar across the bottom of the bulkhead and tied in to the riser bobs on the main bearing.

"Eight thermocouples were placed in the center bulkhead at various locations, as shown in Fig. 1, to study the cooling rates. After these couples were built into the mold, the casting was poured and pyrometer readings started about 4 seconds after pouring.

"The fact that pouring was started at 2650°F., finished at over 2600°F. and that 4 sec. later the first temperature readings were 2000°F., shows an astonishing freezing and cooling rate, as illustrated by Fig. 2.

"Further, the fact that Couples Nos. 1, 3, 5 and 8 were cooling rapidly and Nos. 2, 4, 6 and 7 were heating shortly after readings were started, shows the terrific stressing under way. (See Fig. 1 for couple locations.)

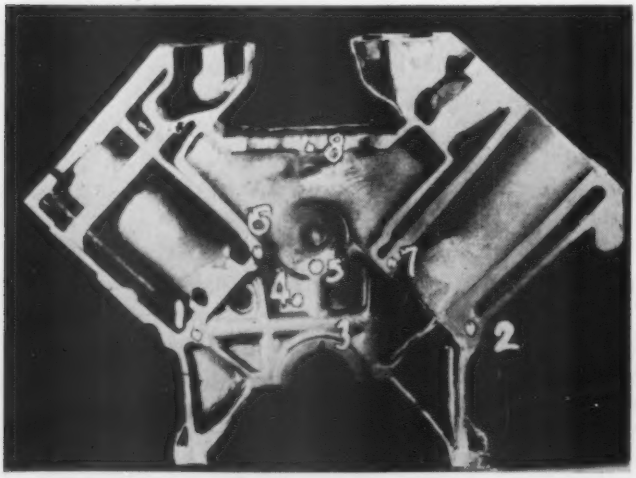

FIG. 1—SECTION OF BLOCK SHOWING POSITIONS OF 8 THERMOCOUPLES.

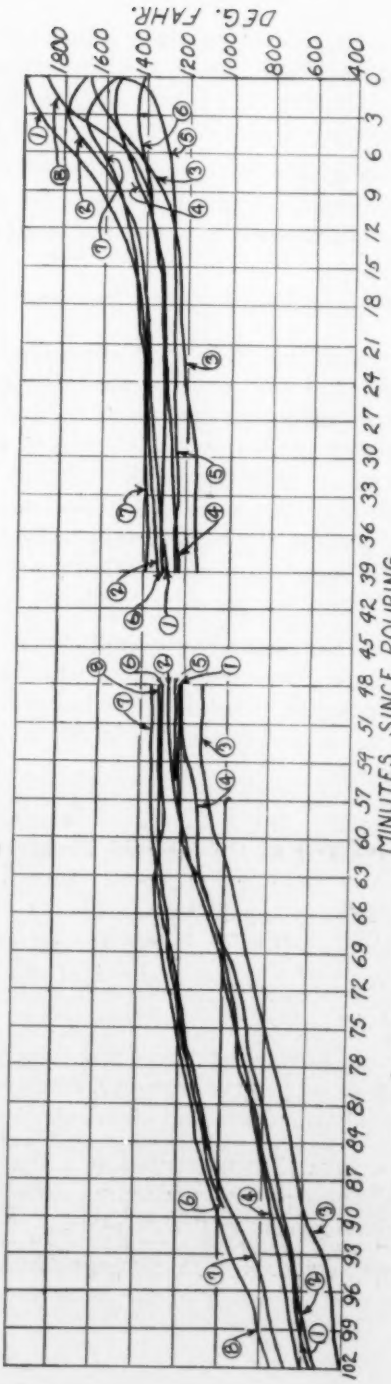

FIG. 2.—COOLING CURVES OF THE METAL AT THE 8 DIFFERENT THERMOCOUPLE STATIONS SHOWN IN FIG. 1.

"Couple No. 5, at the point where the cracking was encountered, was 600°F. below the temperature of No. 1 at 1.5 min. after starting. The terrific cooling rate of No. 5 illustrates graphically why the cracks occurred at this location.

"Also remarkable to note are the differences after removal of core sand. Six hundred forty °F. difference shows why stresses were so high as to rupture castings."

EXAMINING CASTING STRESSES

16. Of interest at this stage is the stress found to exist across the center section. A 1 in. square cross-section cross-tie bar was attached to the two lugs supporting the valve-lifter brackets. Many of these were cracked by the internal stresses in the castings. Not only did these 1 sq. in. cross-ties break while hot, but in some instances they broke after they were cold. Examination of these fractures and tests of similar bars in the tensile testing machine showed the stresses to total approximately 40,000 lb. per sq. in. The iron used gave tensile figures of this order regularly.

17. Other indices of the order and degree of the internal stresses may be gaged from the fact that when a center bulkhead was sawed longitudinally, it broke before the saw cut through and opened up 3/16 in. requiring nearly a 20,000 lb. load on the casting to close the cracks.

18. When cutting end bulkheads, it was necessary to insert wedges behind the saw as the internal stresses caused the block to pinch and break the saw as it went through. Tests made by wedging and re-cutting end bulkheads showed that the saw kerf had to be nearly ⅛ in. wide before the stresses were relieved sufficiently that the kerf would not close tight when the wedges were removed.

19. The engineering department was complaining that cylinder bore warpage in service was excessive, reaching the point where blow-by and loss of power was quite marked.

20. Events from then on shaped to a trend which can best be followed by quoting from subsequent reports. They give the history of the development as it took place.

"It was next desired to determine what difference there was in the rate of cooling of the block with the center cam-

shaft bearing left solid and with risers alongside it. Information also was wanted on how the outside of the casting cooled. Therefore, couples were placed on the cylinder head face and in the oil pan rail.

"This showed a marked improvement in the cooling characteristics of the camshaft bearing. However, primary cooling was still too sharp and rapid in several portions of the block, indicating cold mold conditions; while secondary cooling was much slower than was supposed, indicating need for controlled cooling.

"Twelve couples were placed in the center portion of the mold, as shown in Fig. 3. Couples 1 to 8 were located in the same places as in the preceding test while Nos. 9 and 10 were placed in the oil pan rails and Nos. 11 and 12 in the cylinder head faces over the support struts in the water jackets. Couples 9 and 12 burned off when the casting was poured, but Nos. 10 and 11 gave good indications of rail and head face conditions.

"Note that 1½ min. after pouring at 2600°F., (Fig. 4) a difference of 500°F. was shown between No. 8 couple in the upper part of the water cross pipe and No. 7 couple in lower water jacket wall.

"Note also that 6 or 7 min. after pouring, couple 8, in

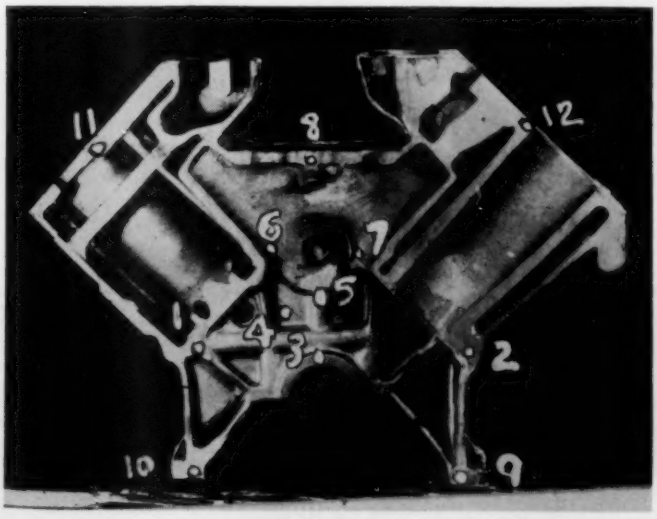

FIG. 3—SECTION OF BLOCK SHOWING POSITION OF 12 THERMOCOUPLES TO BE USED IN SECOND TEST.

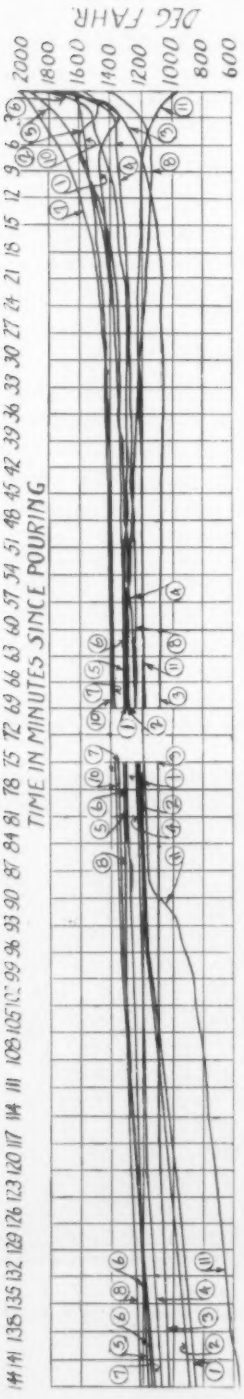

FIG. 4—COOLING CURVES OF THE METAL AT THE THERMOCOUPLE STATIONS SHOWN IN FIG. 8.

upper water cross pipe, and couple 3, in main bearing, are cold compared to the rest of the center section. This called for close study of metal flow and cooling, particularly when it is seen how the added mass at the camshaft bearing, couple 5, has held up the cooling of this portion. Where formerly couple 5 cooled to 1400°F. in one min., it now takes 18 min."

CHANGES MADE AFTER TEST

21. Discussion of each step in the pouring, primary cooling, shakeout, core knockout and secondary cooling, was rife in preparation for the next modification and test. Changes made after this test included adding 86 ft. of cooling conveyor housing, lengthening it to 368 ft. and insulating the walls and roof to hold the heat. This permitted lessening the sharpness of the cooling and cut down the number of cracked castings.

22. The shakeout temperature and time were questioned, so the critical range, Ar $_{3, 2, 1}$ of the iron was checked and found to be 1450° to 1340°F. (Fig. 5).

23. This critical range determination was made on a specimen 4 in. long and 1 in. diameter. On heating this specimen to 1800°F., the total expansion was 0.062 in. or 0.0155 in. per in. of specimen. When cooling from 1800 to 1445°F., the specimen contracted about

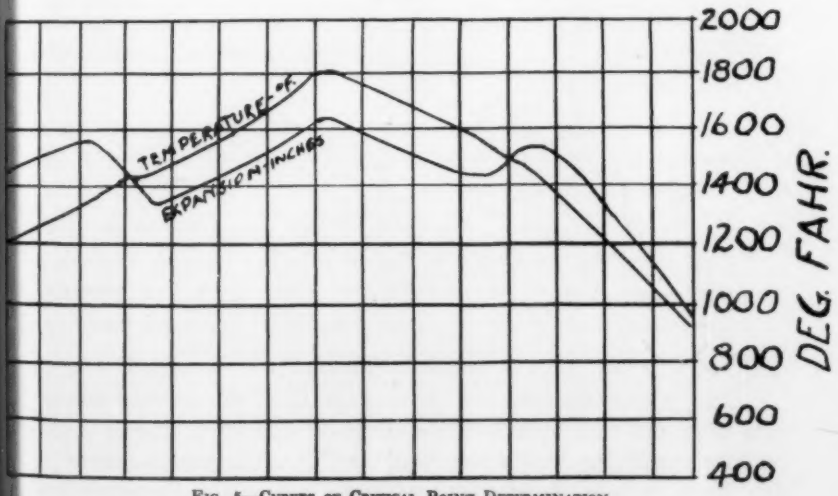

FIG. 5—CURVES OF CRITICAL POINT DETERMINATION.

0.015 in. per in., then expanded approximately 0.011 in. per in. while going through the critical range (at the same time giving off heat so the specimen temperature increased about 5°F.). Thereafter, the specimen cooled and contracted at a uniform rate.

24. The next desire expressed was that we should know what the cooling rate was with the casting shaken out of the mold but with the cores left in position. The suggestion had been made that cores should be knocked out, so as to permit the center of the casting to cool at more nearly the same rate as the outside. The cause of the cracks in the inner water jacket was thought to be the poor heat balance.

25. At this time, the castings were being made with a small hole in the center camshaft bearing to enable the machine shop to start the rough drill for the bearing hole, but without the risers that had formerly been used. This test was the next one run.

26. The report states:

"Primary cooling still indicates that some portions of the mold cool too fast, either from gating or mold temperature conditions.

"Secondary cooling is not at too rapid a rate, in general, but the inside of the casting having the cores left in, does not keep pace with the outside and so sets up extreme stresses which result in cracks.

"Leaving off the risers from the center camshaft bearing has lessened the heat to be dissipated from that portion, but not sufficiently.

"Knocking out the cores immediately after shakeout caused the castings to cool more evenly, thus proving the soundness of this reasoning.

"The proper way to do this job would be to shakeout and knockout the cores and then put the casting through a controlled, heated, cooling kiln or oven. This has been discussed ever since the cracking condition originally came up, but had never had any serious consideration."

27. Since knocking out cores at shakeout time would materially alter the heat balance after shakeout and oven control of the cooling conditions thereafter would affect additional locations in the block, it was decided to use more pyrometer thermocouples

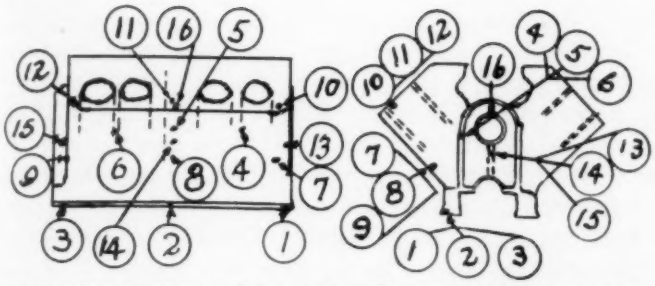

FIG. 6—PLAN OF HOW 16 THERMOCOUPLES WERE TO BE ARRANGED IN CORE.

and get temperature data on other parts of the casting. Instead of the couple locations heretofore utilized, a new arrangement was planned, wherein 16 thermocouples were placed in the cores, along the right hand side of the casting and in the bulkheads, as shown in Figs. 6, 7A, 7B, and 7C.

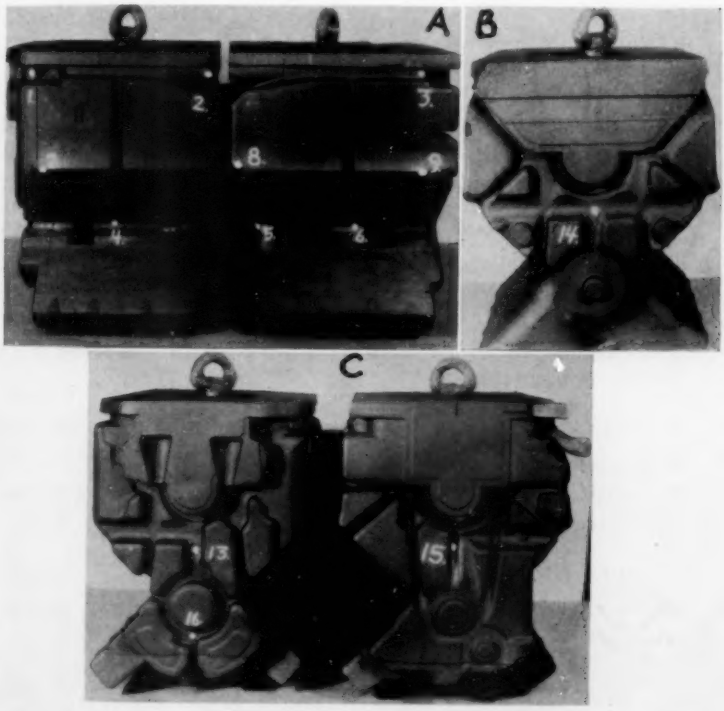

FIG. 7—A, B AND C—POSITION OF THE 16 THERMOCOUPLES IN THE CORES ARE SHOWN BY THE WHITE DOTS AND NUMBERS.

28. The following arrangement was used on all subsequent tests: Couples 1, 2 and 3 were in the oil pan rail; front, center and rear. Couples 4, 5 and 6 were in the inner water jacket wall, where the side slab joins the bottom slab; at the center of the front section, at the junction of the cross water passage, and at the center of the rear section. Couples 7, 8 and 9 were in the outer wall, where the oil pan wall, outer jacket wall and bottom slab join; front, center and rear. Couples 10, 11 and 12 were in the top slab (cylinder head face), midway from bore center line to outside edge; front, center and rear. Couples 13, 14 and 15 were in the bulkheads, in the oil tube ribs between the main and camshaft bearings; front, center and rear. Couple 16 was in the center bulkhead, in the top of the camshaft bearing.

29. To quote further from the report:

"It was also decided to torch-heat the drags on the conveyor, in order to improve heat balance conditions, so this was included in the test.

"Pouring was at 2760°F. and within one min. a difference of 1100°F. was shown between the center camshaft bearing and the rear head face. A difference of 1000°F. showed between the center camshaft and the rear outside wall. (Fig. 8).

"The cooling from pouring temperature to 1400°F. took about 30 min. showing that the practice of torch heating the drag has improved the primary cooling by slowing it down.

"It was suggested that other portions of the mold might be heated to advantage.

"The cooling curves show that the conditions are such as to allow the outside of the block to cool, contract, go through the critical range with attendant expansion and again contract—before the inside reaches the critical range.

"To minimize this, the temperature differences between adjacent portions of the block should be cut down and the cooling rate more evenly controlled.

"By proper control, the cooling time could be actually cut down, while still reducing internal stresses and obviating cracks."

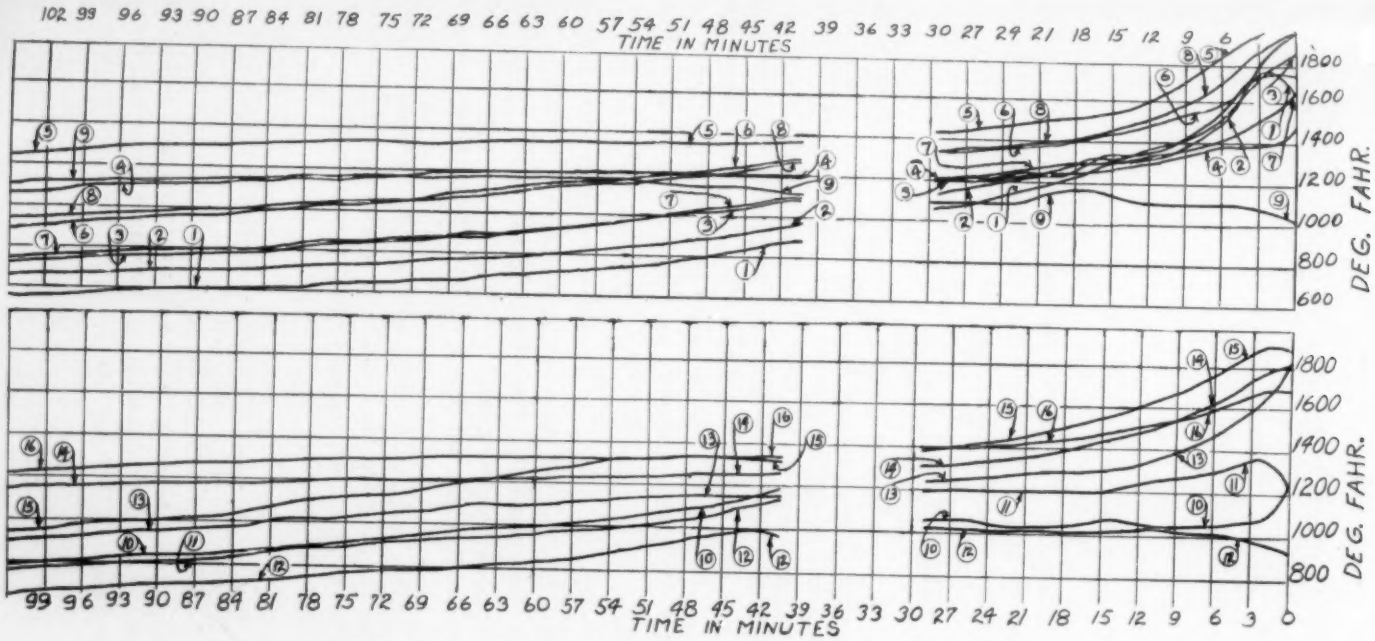

FIG. 8—COOLING CURVES OF THE 16 THERMOCOUPLES SHOWN IN FIG. 7A, B AND C.

TRY WATER COOLING

30. By this time, it was agreed fairly well among the foundry staff that a cooling oven on the carryoff conveyor would be decidedly advantageous. However, the cost looked high and so several digressions were undertaken in the effort to get results at lower cost.

31. One proposal was that, after pouring, water should be added to the hollow body cores to cool off the center mass. This was greeted with some trepidation, as everyone visualized explosions, wrecks and a generally impractical operation. Despite the fears, the test was made. The water did what it was calculated to do. It took considerable heat out of the heavy center mass.

32. The report continues:

> "In figuring the amount of water to be added, it was estimated (from study of the chart of the last cooling test) that a drop of 300°F. of about 30 lb. of iron in the center of the block would be all right to try out first. Assuming a specific heat of 0.12 for cast iron, gave 1080 B.T.U. to be extracted. With service water at 60-65°F., we could depend upon a rise of 150°F. (from the water temperature to the boiling point). This showed slightly over 7 lb. of water, and to play safe, 3 qt. was decided upon.
>
> "The charts (Fig. 9) for this test showed that a temperature drop could be produced by adding water, but that even when this was done, the outer portions of the block cooled faster than the inside and that the center portion which had been water cooled was taking up heat from adjacent masses."

33. After this first trial of adding water, several additional tests along the same line were made. All confirmed the opinion that shakeout time should be studied and that cores should be knocked out at shakeout time.

CHANGE GATING ARRANGEMENT

34. Changes were made in the gating (Fig. 10) by blocking out several gates along each side near the center and adding them along each end. These modifications altered and improved the cooling characteristics, but not enough to fix the job. Changes were

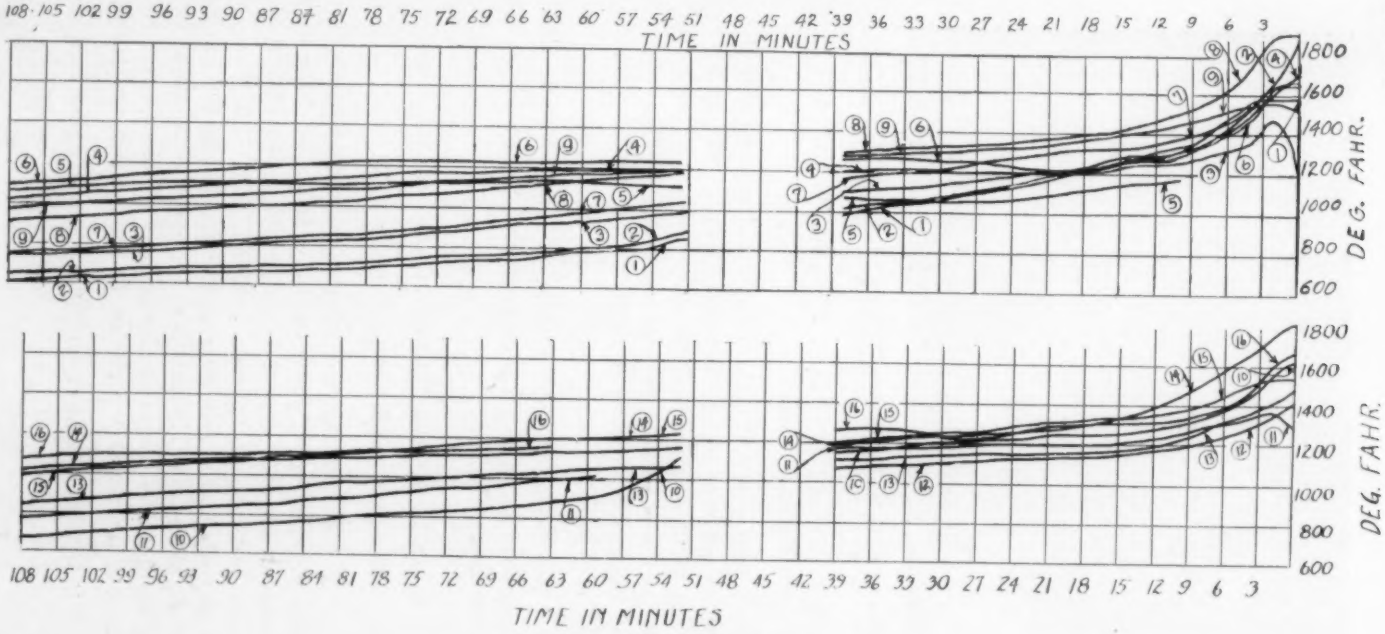

FIG. 9—RESULTS OBTAINED BY WATER COOLING.

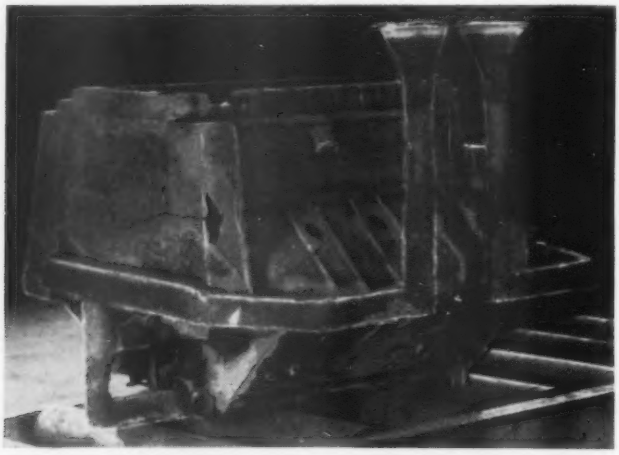

FIG. 10—CASTING IN WHICH CHANGES WERE MADE IN THE GATING BY BLOCKING OUT SEVERAL GATES ALONG EACH SIDE NEAR THE CENTER AND ADDING THEM ALONG EACH END.

also made in breakoff and risers. These also improved the job, but still left something lacking.

35. A test was run with the modified gating, with the cores all knocked out and the casting placed in a welding pre-heating furnace at 1100°F., after shakeout and knockout. This showed real promise, as the controlled cooling indicated that the various portions of the casting cooled at more nearly similar temperatures. (Fig. 11).

36. The conclusions so far seem to build up to the fact that there was need for a carryoff cooling oven to control the cooling of the castings.

MAKE METALLOGRAPHIC AND CHEMICAL ANALYSES

37. The report continues as follows:

"Before proceeding further, it was decided that a complete metallographic and chemical examination be given three blocks which had cracked in the center bulkhead between the main bearing and the cam bearing (Fig. 12). These blocks were cast on the following dates: Block No. 100 on May 4, 1936; blocks No. 110 and No. 111 on May 5, 1936. Samples of bores, taken from opposite corners of each block, also were checked for possible variation in composition or structure.

"All photomicrographs indicate that the iron used in these

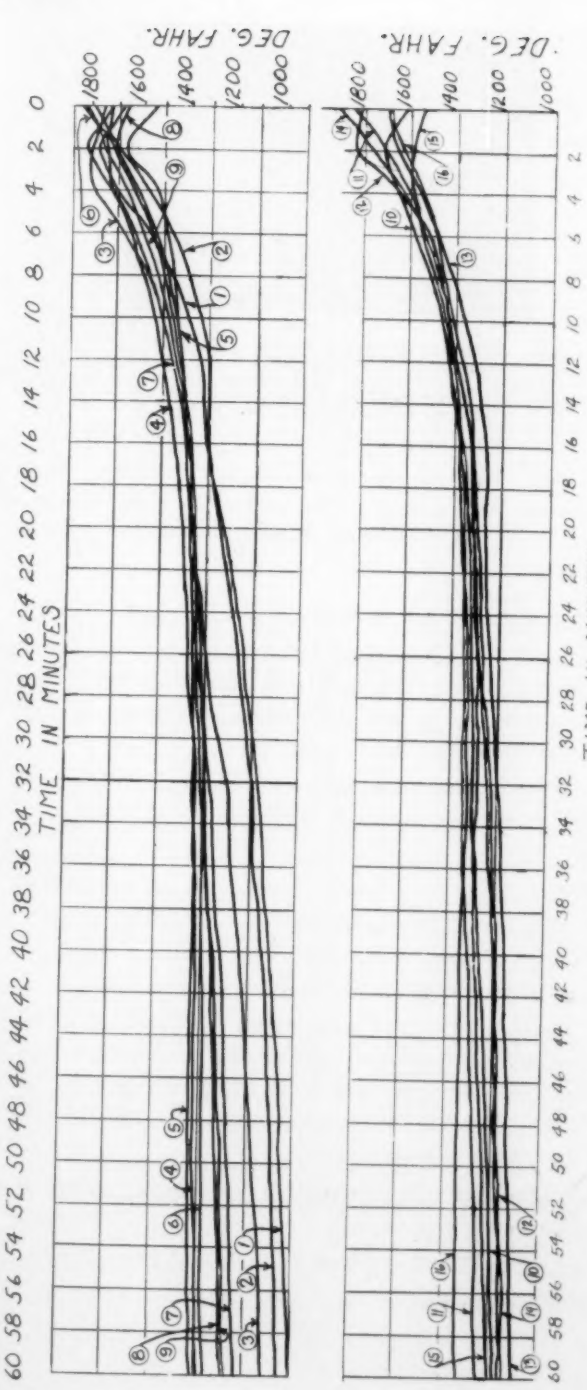

FIG. 11—RESULTS OF PLACING CASTING IN PRE-HEATING FURNACE.

FIG. 12—PHOTOGRAPH OF CENTER BULKHEAD WHICH CRACKED FROM MAIN BEARING TO CAMSHAFT BEARING THROUGH OIL HOLE. (SAMPLE 2058, BLOCK NO. 100, CAST MAY 4, 1936). THIS ILLUSTRATES POSITION AND DIRECTION OF FRACTURE. OUTLINES SHOW POSITIONS OF MICRO SAMPLES. CHIPS FOR ANALYSIS WERE TAKEN ALONG FRACTURE AFTER MICROS WERE TAKEN. (APPROXIMATELY HALF SIZE.)

blocks was a good, average grade of pearlitic iron, and in some cases, better than average. The variations noted in the microstructures can be attributed directly to cooling conditions in different sections.

"Fractures were all medium to fine grain.

"The difference in hardness between the cam bearing, the main bearing, and the bulkhead section between them can be accounted for by the differences in cooling rate at these points due to variations in sections.

"Chemical compositions varied slightly within each block but the variations were so slight as to be relatively unimportant.

"The cracking of these blocks cannot be accounted for either metallurgically or chemically. Breakage is due to mechanical or internal stress occurring during cooling. The remedy lies in modification of the foundry procedure.

"Photographs were taken of all fractures, chips for chemical analysis were taken at the points of fracture, and photomicrographs were obtained from samples taken at the points of fracture. Brinell hardness readings were taken on all bores, cam, and main bearings.

"All pieces were designated by letters according to the following code:

A and A1—The two pieces of bulkhead from Block No. 111, Sample No. 2060.

B and B1—The two pieces of bulkhead from Block No. 110, Sample No. 2059.

C and C1—The two pieces of bulkhead from Block No. 100, Sample No. 2058.

D—Sample of bore, left bank rear, from Block No. 111.

D1—Sample of bore, right bank front, from Block No. 111.

E—Sample of bore, right bank front, from Block No. 110.

E1—Sample of bore, left bank rear, from Block No. 110.

F—Sample of bore, right bank front, from Block No. 100.

F1—Sample of bore, left bank rear, from Block No. 100.

"Block No. 100, sample 2058, was selected as typical and the data of Table 1 illustrates the hardnesses and compositions."

Table 1

TYPICAL SAMPLE STUDIED TO DETERMINE CAUSE OF CRACKS

Sample No. 2058—Block No. 100—Cast May 4, 1936

BRINELL HARDNESS

Bulkhead		*Right Bank Front Barrel*	*Left Bank Rear Barrel*
C	C1	F	F1
Cam 171	Cam 171	182	182
Main 194	Main 194		

CHEMICAL COMPOSITION, PERCENT

	C	C1	F	F1
Total Carbon	3.220	3.140	3.140	3.140
Combined Carbon	0.790	0.680	0.410	0.420
Graphitic Carbon	2.430	2.460	2.730	2.720
Manganese	0.530	0.540	0.580	0.570
Phosphorus	0.146	0.147	0.161	0.157
Sulphur	0.079	0.078	0.090	0.089
Nickel	Nil	Nil	Nil	Nil
Chromium	0.040	0.040	0.040	0.040
Silicon	2.240	2.240	2.150	2.240
Copper	0.470	0.550	0.480	0.510

"These structures (Figs. 13 to 21) and compositions are representative of what had been examined from scores of previous cracked blocks and are typical of the melting practice at that time."

INSTALL CONTROLLED COOLING CYCLE

38. As will be noted, many of these blocks cracked after the vertical cam-to-crank oil holes were drilled. This might mean an elapsed time of from one day to three weeks after casting.

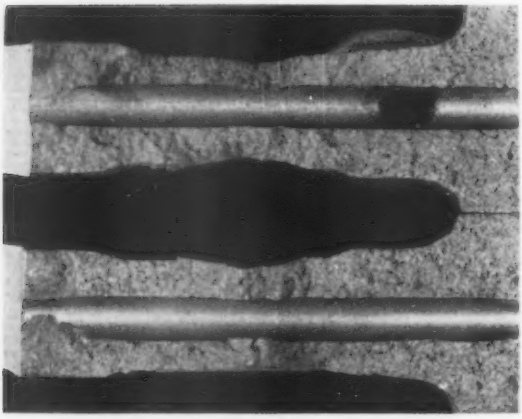

FIG. 13—PHOTO SHOWING THE FRACTURE OF THE BULKHEAD. 1 X.

39. When, finally, it became apparent that a controlled cooling cycle was essential, the project was taken up with the equipment people. One after another of several reputable firms engaged in building such equipment refused to bid on the proposition. They said it was neither oven, furnace, nor kiln, and would not consider it because it involved what they termed a "trick process," so they would have none of it.

40. Finally, with the co-operation of the maintenance and construction department, one of the more venturesome furnace builders installed the jobs. When it was set up, it appeared to be

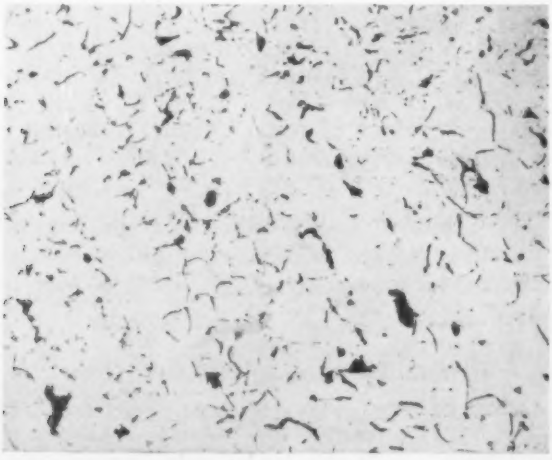

FIG. 14—MICRO TAKEN AT MAIN BEARING. BULKHEAD CI. 100 X. UNETCHED.

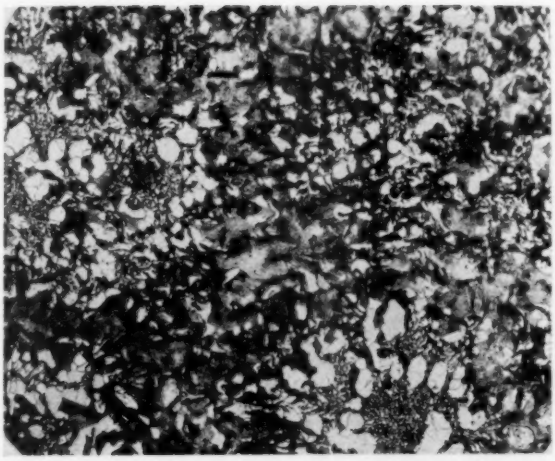

FIG. 15—MICRO TAKEN AT OUTER SURFACE OF CENTER SECTION. BULKHEAD C. 100 X. ETCHED 4 PER CENT NITAL.

of no avail, as several blocks cracked in the center bulkhead on the first day's operation of the oven; so it was desired to know whether the oven operating temperatures were according to the specifications laid down before the oven was designed and installed.

41. The report on this test was as follows:

"The entering and soaking portion of the oven (the first 30 min.) was supposed to hold at 1100°F. so as to equalize the

FIG. 16—MICRO TAKEN AT HEART OF CENTER SECTION. BULKHEAD C. 100 X. ETCHED 4 PER CENT NITAL.

FIG. 17—MICRO TAKEN AT HEART OF CENTER SECTION, BULKHEAD C. 500 X. ETCHED 4 PER CENT NITAL.

FIG. 18—FRACTURE OF CYLINDER BARREL. 2 X. UNETCHED.

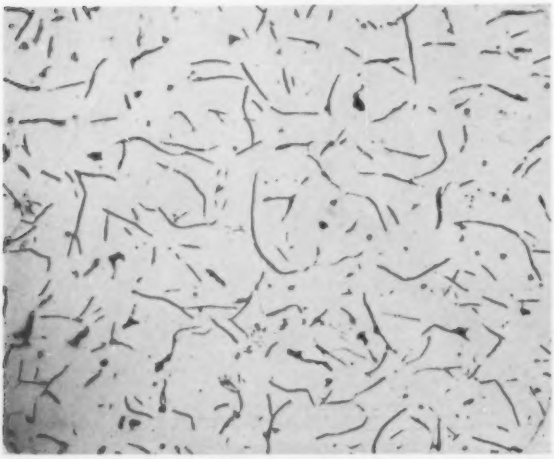

FIG. 19—MICRO OF CYLINDER BARREL FI. 100 X. UNETCHED.

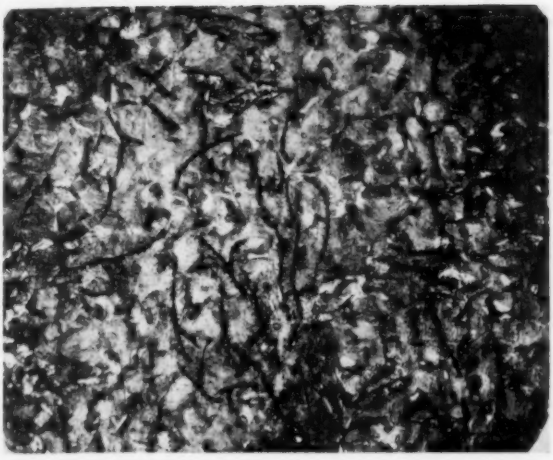

FIG. 20—MICRO OF CYLINDER BARREL FL. 100 X. ETCHED WITH 4 PER CENT NITAL.

temperature of the blocks, which might be charged at temperatures from 900 to 1400°F.

"The cooling portion of the oven was supposed to cool the block evenly to below 500°F., preferably 400°F., over a 2½ hr. period.

"As operating the first day, the castings enter the oven at about 1100°F., but cool off rapidly (in 5 min.), to 860°F. (Fig. 22); then heat up again to 960°F. (for 13 min.), after

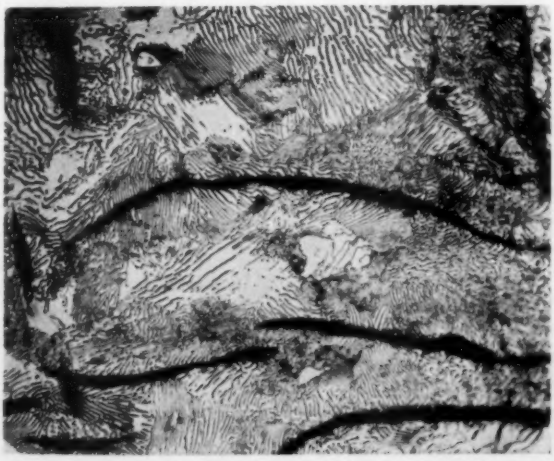

FIG. 21—MICRO OF CYLINDER BARREL FL. 500 X. ETCHED 4 PER CENT NITAL.

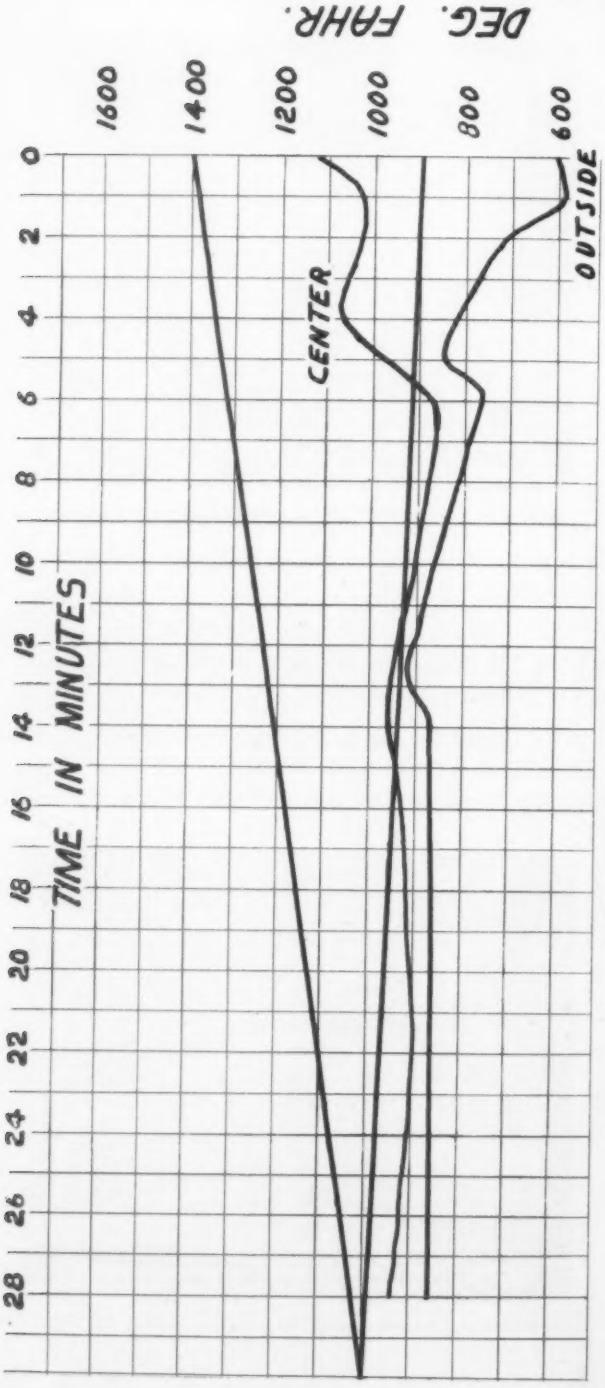

FIG. 22—FIRST CURVES OF CONTROLLED COOLING CYCLE INSTALLATION.

which they remain at nearly the same heat for some time (940°F., for 28 min.). The oven control temperature is 860, instead of 1100°F. At the discharge end, the blocks come out at 630 to 680°F., instead of below 500°F.

"In making this test, two trailing thermocouples were attached to a block casting as it was hung on the conveyor; one on the center bulkhead and the other on the outside of the casting. These were allowed to follow the casting for 28 min. as it entered the oven. Temperatures were read every minute on the trailing couples and three times on the oven control pyrometer. The conveyor speed was 40 in. per min.

"The oven control should be raised up to 1050 to 1100°F. There should be circulation of air in the cooling portion of the oven, so that the inside and outside temperatures of the castings will be more uniform. This circulation should be controlled so as to get the block out at below 500°F., preferably 400°F.

"The conveyor speed should be reduced to where the cooling time will be 2½ hr."

42. The construction division did some remodelling, as a result of this report, after which the job went along in good shape until the next spring when several cracked blocks were encountered. At this time, it was agreed that the oven cooling curve should be checked.

43. The check test showed the following:

"The first test (Fig. 23) showed temperatures considerably low, with too rapid cooling during the early part of the run through the oven. (The straight line is the specified heat gradient). Some open heater connection holes were closed up and another test was run (Fig. 24), which showed good results.

"The conveyor was slowed down to 24 in. per min. which was stated to be equal to 45 blocks per hour. It had previously been running much faster.

"This test indicated that automatic controls should be installed on the burners so as to maintain temperatures more uniformly.

"It seemed possible that in cold weather, additional heat would be needed at the entrance of the oven.

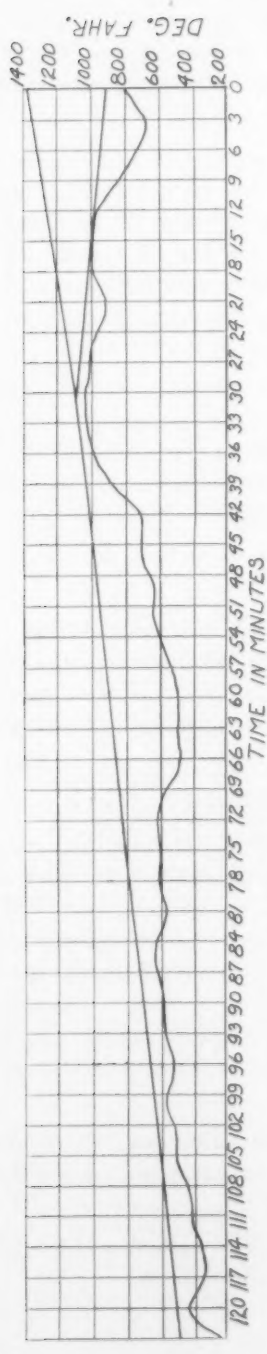

FIG. 23—CURVES OF THE TEST ON OVEN COOLING CURVES.

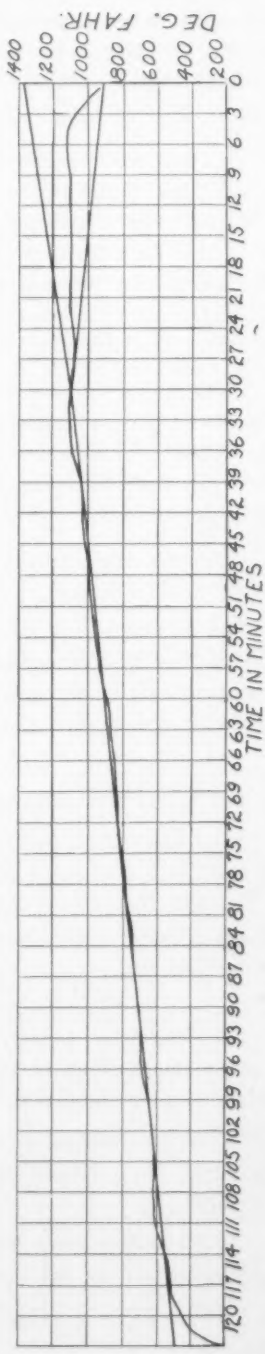

FIG. 24—TEST CURVES SHOWING IMPROVED RESULTS AFTER HEATER CONNECTION HOLES WERE CLOSED UP.

"Tests were then run at discharge of oven, to assure blocks coming out below a dangerous heat. It was agreed, that, if blocks came out too hot, an addition to the discharge end of the oven, made up of sheet metal housing only, should be installed.

"In these tests, one long trailing couple was fastened to the center bulkhead of a block after core knockout. A portable indicator was connected and readings taken every minute as the conveyor carried the block through the oven."

44. These findings called for some rebuilding of the conveyor and oven top, where the hook slot allowed too much heat loss.

COOLING OVEN REBUILT

45. After rebuilding the oven, it was desired to check the temperature curve with a trailing thermocouple to ascertain what the operating conditions were.

46. The report for this test read:

"The time to equalize is fairly good, but the equalizing temperature is low, as is.

"However, the first cooling is bad in that there is a 300°F. drop in 12 min. This is far too abrupt and should be lessened.

"There is another even more severe drop at 50 min., where the temperature falls 280°F. in 5 min.

"In the test (Fig. 23), the operating curve approximates the specified gradient. The peaks and dips in the curve of Fig. 25 should be eliminated by oven and burner adjustments so as to give as good a chart as the old oven formerly did.

"The crooked line shows the readings as taken, and is indicative of the temperature of thin sections on outside of block casting. The evenly curved line has been inserted by hand, from the crooked line readings, to show the probable temperature trend of the heavy mass. The straight line shows the specified curve, as noted on the layout blueprint.

"The first abrupt drop, at 30 min., is at the slope down from the upper to the lower level; after the hot zone.

"The second, at 50 min., is approximately at the slope up from the lower to the upper level.

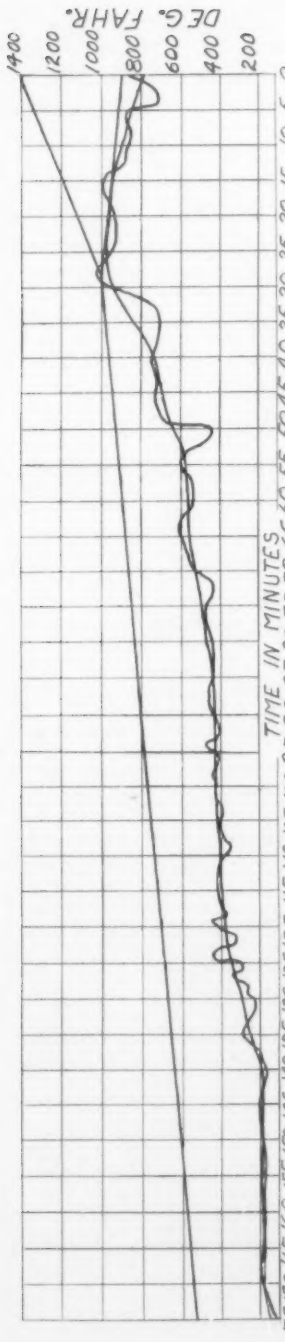

FIG. 25—TEMPERATURE CHECK CURVES TO ASCERTAIN WHAT THE OPERATING CONDITIONS WERE.

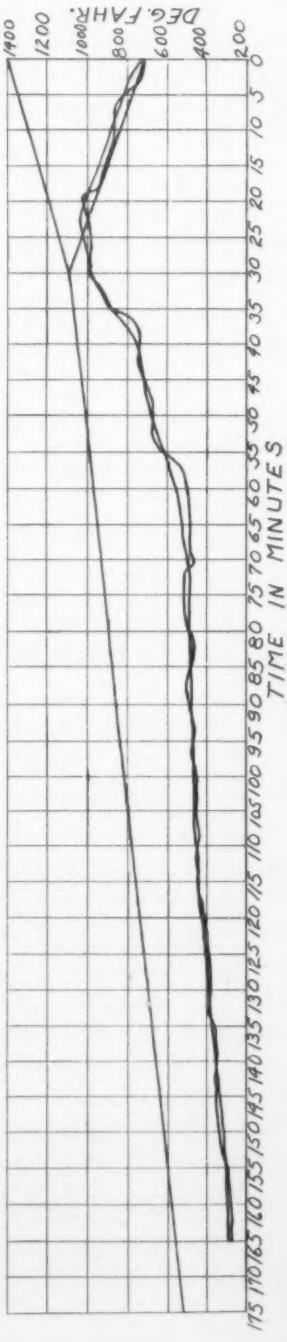

FIG. 26—RESULT OF RETEST.

"There must be a draft to cause such abrupt drops in the temperature at the points noted. There naturally would be a marked change at the first slope down due to thermal air circulation, the heated air tending to rise and the cold air to fall. However, the shape of the curve shows that more than this is at work to cause so large a drop."

47. After this experiment, the trouble was aggravated instead of lessened, so another re-vamping was discussed. For some time no action was taken, as production went along fairly well. However, in the winter, as cracked blocks were again encountered, it was decided to run another test to ascertain the cause.

RESULTS OF RETEST

48. The results of the test as reported are as follows:

"The time to equalize is short and the equalization temperature low. First cooling is bad as during the interval from 32 min. to 42 min. the temperature drops about 300°F. Cooling from 42 min. to 75 min. is still too rapid, the drop being about 230°F.

"The oven conveyor was running at a speed of 20 in. per min. which is not too fast, but the heat is low and the equalizing time is too short.

"The curve, Fig. 26, like that of Fig. 25, is not so good as the one of Fig. 23. Oven and burner adjustment should be made to bring the operating curve up to the specified rate.

"Setup was similar to that of Figs. 23, 24 and 25. Charting also similar giving the following: Crooked line—instrument readings; evenly curved line—probable block temperature; straight line—specified curve.

"The first abrupt drop, 32 to 42 min., occurs at the down slope from the upper level of the oven to the lower or floor level. The second drop, 42 to 75 min., is along the lower level; to the beginning of the up slope to the second upper level. From here on the cooling is uniform and gradual, but the temperature has already gotten too low. The end of the curve, at 167 min., was not at the discharge of the tunnel, but about 60 ft. back from the discharge end of the housing."

49. After this revision, the job went along much better and no scrap due to cracks was noted during the remainder of the production season.

NEW DESIGN CRACKS

50. Prior to the start of the next production season, a change in design was made. A ventilating pipe, down through the flywheel end of the block, was added. This stiffened the right rear corner of the block and, as a result, cracks began to appear in the right hand inner water jacket wall, alongside the cross water passage.

51. Another change which affected the operation was that the center body cores were changed from dry sand to green sand, thus introducing additional variables not hitherto encountered.

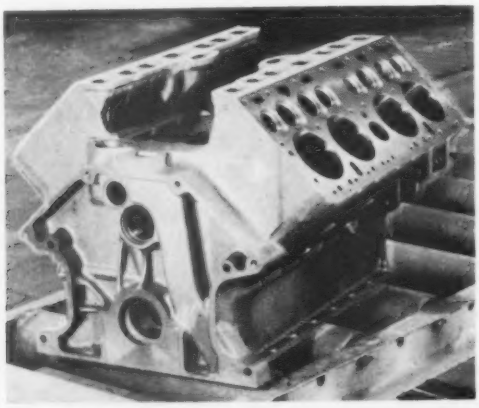

FIG. 27—PHOTO OF THE 1939 BLOCK IN WHICH CRACKING WAS ENCOUNTERED.

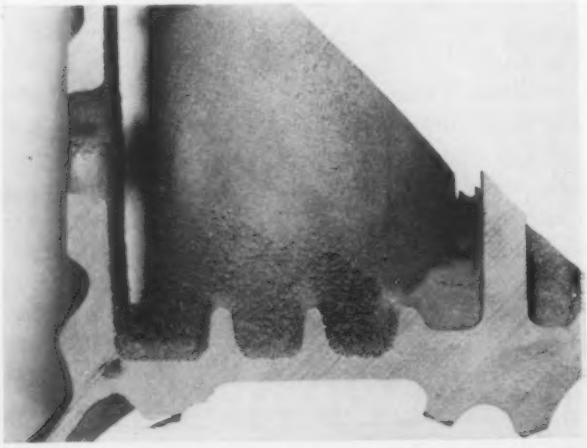

FIG. 28—SECTION SHOWING THICK SLAB BETWEEN MIDDLE BARRELS GIVING BETTER HEAT BALANCE.

52. Note (Fig. 27) that the inner water jacket wall on the left side had an offset in it which makes it more flexible, while on the right side where the vent pipe stiffens the corner, the inner water jacket is straight, with no possibility of flexing to relieve the longitudinal stress.

53. The cracking persisted even with good carryoff conveyor oven control. Check tests were run on the oven and showed good heat gradient—almost like the ideal specified curve.

54. Several remedies for the trouble were tried and one finally was suggested which worked satisfactorily. This was to thicken the bottom cylinder barrel slab between the two center barrels, on the right side (Fig. 28). This additional mass apparently changed the heat balance of the structure and reduced the local stressing which had caused the cracking.

55. Here again is evidence of the difference between heat balance and heat unbalance. Evidence of the latter is obtainable by the methods indicated above and means for changing from unbalance to balance will suggest themselves depending upon the structure being dealt with.

DISCUSSION

Presiding: F. J. WALLS, International Nickel Co., Detroit, Mich.

In absence of the author, this paper was presented by H. B. Swan, foundry superintendent, Cadillac Motor Car Co., Detroit, Mich.

F. E. REINERS:[1] Did the oven treatment have any effect on the machinability of the castings?

MR. SWAN: The highest temperature is only 1100°F. and it had absolutely no effect upon the casting other than equalizing the temperature and cooling the casting. It is not an annealing operation.

CHAIRMAN WALLS: Mr. Swan, did you have less dimensional distortion after machining than you did prior to the installation of the oven?

MR. SWAN: Yes, there is no question about that. We had had some complaint from the engineers on distortion in the bore but after this operation was installed, we did not hear any more about that.

MEMBER: Were the cracks discussed in this paper cold cracks or hot cracks?

MR. SWAN: Both. There were times when we would go along the cooling line in the morning, when the blocks were, about as near as we

[1] York Ice Machinery Corp., York, Pa.

could judge, about 700°F. and we could hear them crack. There were other times when the cracks would not develop possibly for 2 or 3 days later in the machine shop. With a casting under stress, the machining sometimes relieves certain points and the casting would crack in the machine shop.

I think Mr. Danse was persuaded to write this paper, even though it was about 3 years after the trouble occurred, because he felt that this means of study, the use of thermocouples in the casting, was a tool that could be used to advantage by the foundry industry, perhaps not always for cracks, but sometimes for warpage conditions.

A. HIERTGEN:[2] On the cooling curves shown, I did not understand how the heating rate of the casting would continue at a uniform rate after the shake-out. I noticed on the curves, that following the shake-out period, the curve showed a uniform increase in temperature after the shake-out.

MR. SWAN: Mr. Danse mentioned that in his paper, I believe. He felt that some of the curves were hard to explain. The reason that there was an increase in temperature, was due to heat absorption from other parts of the casting and also, from the fact there is extreme heat absorption around the barrels from the jacket sand. There were a lot of points that we were not always able to explain, but these were the facts as we found them.

[2] Henry Vogt Machine Co., Louisville, Ky.

High Strength Cast Steels*

By Georges Delbart,† Denain, France
Translation by W. H. Spencer,** Muskegon, Mich.

Abstract

High strength cast steels are discussed from a metallurgical viewpoint. The author divides his presentation into the following headings: Defects of cast steel, structure of cast steel, refinement of secondary and tertiary structures, hardening, French specifications and definitions. Under defects of cast steels he discusses interdendritic shrinkage or microshrinkage. Primary structure and homogenization are discussed under structure of cast steel while granulation and hardening are discussed under refinement of secondary and tertiary structures. In the subject of French specifications, he presents specifications as required by the A.F.N.O.R. Micros and drawing are included to better illustrate the various points discussed.

Introduction

1. The problem of light weight in construction has found numerous solutions in the utilization of special cast steels. Decreases in weight of 30 to 35 per cent and even of 50 per cent are cited readily.

2. The steel foundry has followed metallurgical progress with a certain amount of delay. The foundry can not adopt the findings from forgings and rolled material because sometimes cast steel retains its solidification structure in spite of thermal treatment. Moreover, ingots for forging are cast under the best conditions and undergo profound transformations and improvements while being worked. We wish to present here the facts that we believe essential to the metallurgical practice of the steel foundry.

3. A casting should, above all, conform to the drawing and this depends on the shrinkage and flowability of the metal employed. The question of flowability was the subject of a paper by Portevin[1] in 1932. A casting of high quality should be free of defects and have a fine structure.

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*Official Exchange Paper, presented on behalf of the Association Technique de Fonderie de France.

†Chief Engineer, Metallurgical Department, Cail.
**Metallurgist, Sealed Power Corporation.

Note: Presented Before Steel Session of the 43rd Annual A.F.A. Convention, Cincinnati, Ohio, May 17, 1939.

DEFECTS OF CAST STEEL

4. We have previously considered some of the defects of cast steel, such as blow holes, inclusions, and cracks[2]. These defects have been combatted for a long time and are beginning to be well understood. In attempting to obtain higher qualities, attention was turned to the defects whose importance was secondary for ordinary cast steel, but primary in the search for higher mechanical properties. These latter defects were dependent less on the specific properties of the metal than on its composition. There appeared in particular the interdendritic shrinkage or microshrinkage, which was first reported in steel ingots by Charpy in 1920. Interdendritic shrinkage in cast steel was the subject of a recent article by Portevin[3]. We repeat its essentials in the following paragraphs.

5. Microshrinkage appears: (a) in part, in regions of major segregation during the second stage of solidification by flowing off of the liquid metal through the interlaced dendrites of solidified metal; (b) and especially during the third stage of solidification, as a consequence of the shrinkage of the last liquid portions imprisoned in the dendritic cells (Fig. 6B).

6. The region of "basaltic" structure which is found in all ingots poured in iron ingot molds does not contain microshrinkage because of the regular progression of the crystallization, but this structure practically does not exist in castings poured in sand because of their slower solidification.

7. The formation of microshrinkage depends on the specific properties of metal and on the solidification conditions and, consequently, on the pouring temperature and cooling rate.

8. It is necessary, in order to attain high mechanical strength in a cast metal, that a fine structure be obtained in addition to great compactness. This condition can not be discussed without a preliminary agreement on definitions of structure.

STRUCTURE OF CAST STEEL

9. We have previously discussed the primary and secondary structure of cast steel[4], the primary structure being the structure of solidification (Figs. 1, 2, 7, 8A, 8B). In order to avoid all con-

[1]World Foundry Congress in Paris, Bulletin de l'Association Technique de Fonderie, Sept., 1932.
[2]Bulletin de l'Association Technique de Fonderie, Dec., 1934.
[3]Metal Progress, Aug., 1938, p. 152.

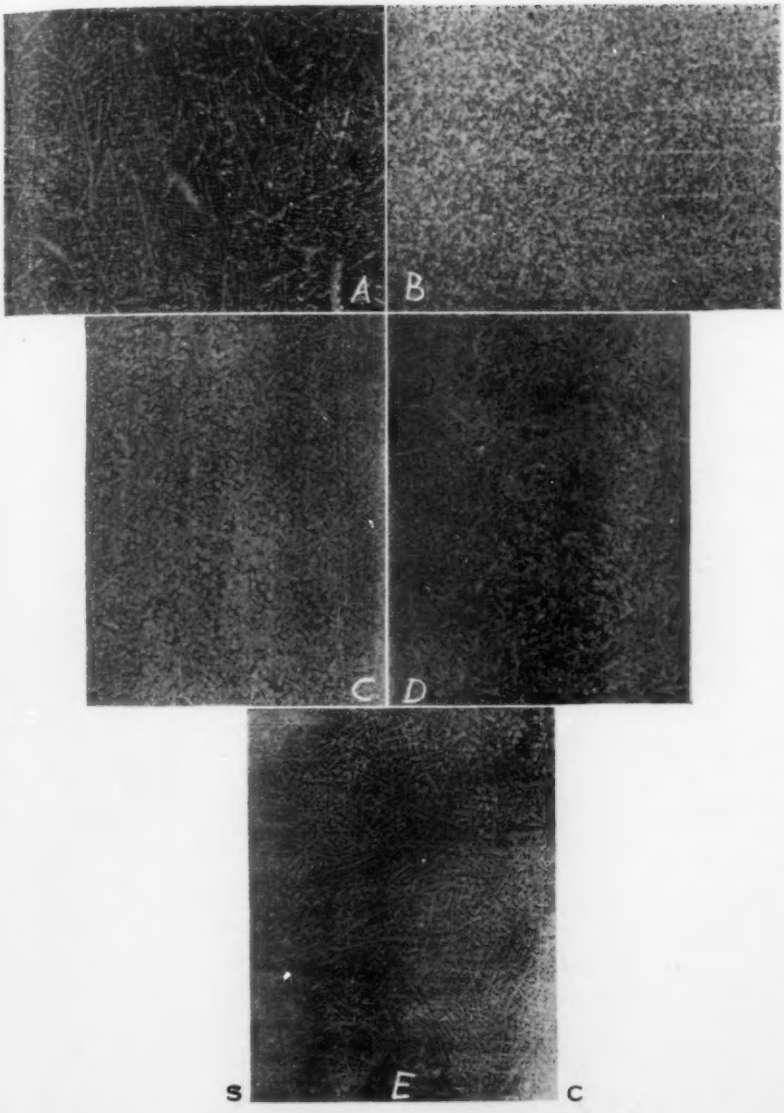

FIG. 1—PRIMARY STRUCTURE, NATURAL SIZE, OF DIFFERENT SPECIAL STEELS. (A) NI-CR-MO STEEL. (B) NI-CR-MO-V STEEL. (C) CR-MO-V STEEL. (D) CR-MO STEEL. (E) CR-MO STEEL. THE WIDTH OF EACH PHOTOGRAPH REPRESENTS THE TOTAL THICKNESS OF THE CASTING FOR A, B, C, D, BUT ONLY A PORTION OF THE THICKNESS FOR E. IN THE LATTER, S INDICATES THE SURFACE AND C THE CENTER OF THE CASTING. MICROS D AND E, STEEL OF THE SAME COMPOSITION, SHOW A COARSER STRUCTURE IN THE THICKER CASTING E WHICH COOLED MORE SLOWLY.

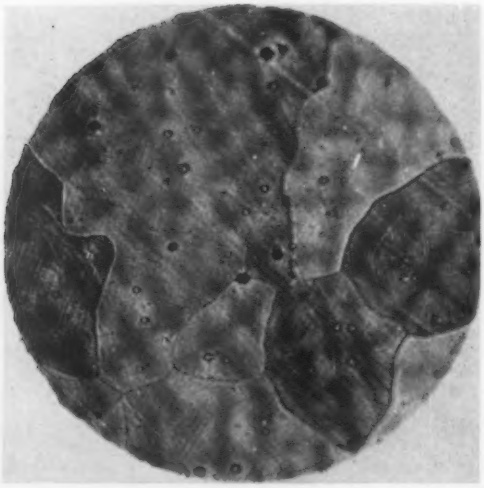

FIG. 2—PRIMARY STRUCTURES IN A 12 PER CENT MN STEEL, EACH GRAIN OF AUSTENITE CONTAINS A CHARACTERISTIC DENDRITIC SKELETON. x200.

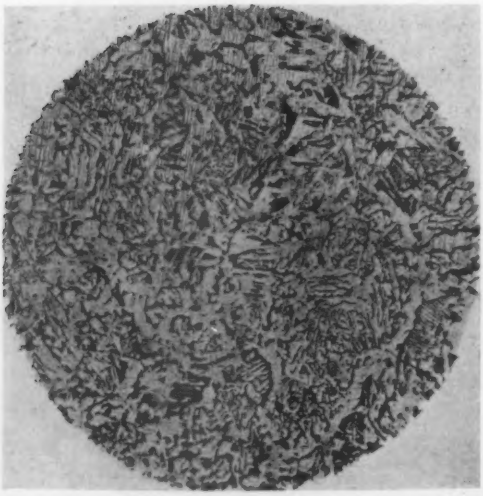

FIG. 3—SECONDARY STRUCTURE (FERRITE IN NETWORK AND IN GRAINS) AND TERTIARY STRUCTURE (ACICULAR PEARLITE) IN A 0.20 PER CENT C CAST STEEL (AS CAST). x50.

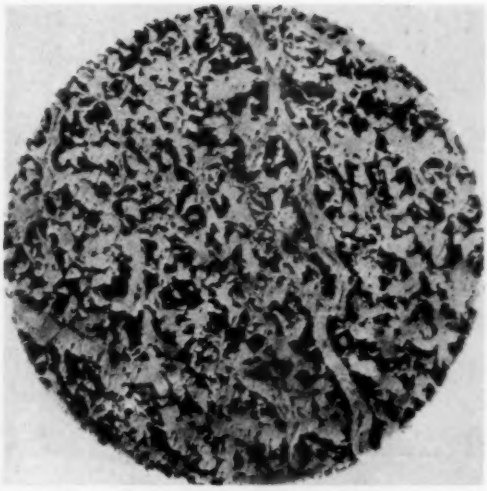

FIG. 4—SECONDARY STRUCTURE (FERRITE IN NETWORK AND IN GRAINS) AND TERTIARY STRUCTURE (PEARLITE IN GRAINS) IN A 0.80 PER CENT C CAST STEEL. (ANNEALED AND COOLED SLOWLY). x100.

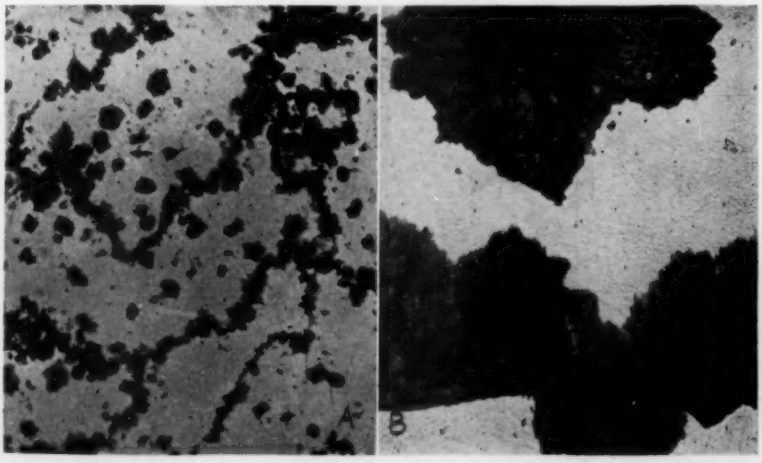

FIG. 5—(A) SECONDARY STRUCTURE IN A CAST HARDENING STEEL DISTINGUISHED BY A TROOSTITE NETWORK IN A BACKGROUND OF MARTENSITE. (AS CAST). x25. (B) SAME STEEL AS IN *A* SHOWN UNDER HIGHER MAGNIFICATION. x400.

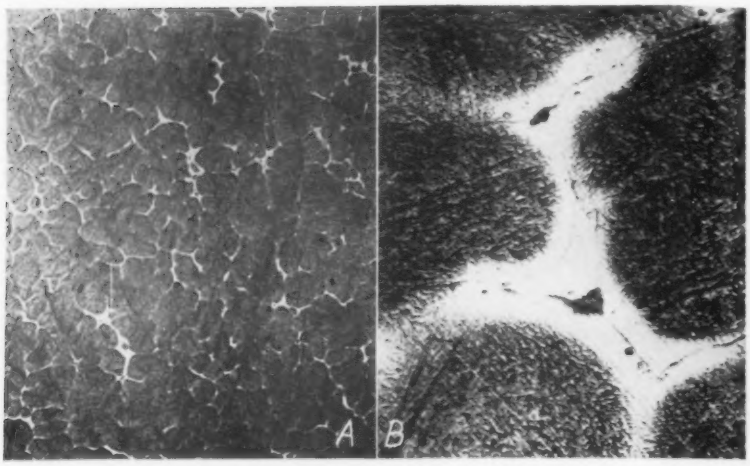

FIG. 6—(A) SECONDARY STRUCTURE IN A CAST HARDENING STEEL DISTINGUISHED BY A FERRITE NETWORK IN A BACKGROUND OF MARTENSITE. (AS CAST). X25. (B) SAME STEEL AS IN *A* SHOWN UNDER HIGHER MAGNIFICATION. THE FERRITE IS COLLECTED AT THE GRAIN BOUNDARIES AND CONTAINS INCLUSIONS AND MICROSHRINKS. X400.

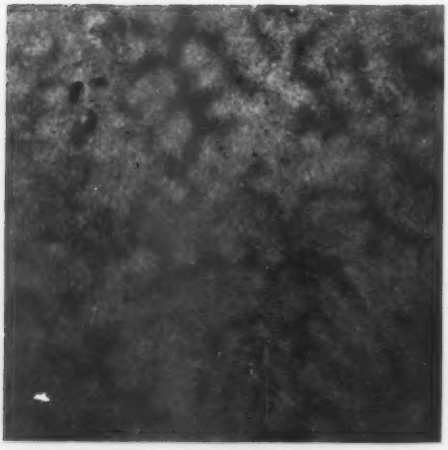

FIG. 7—STRUCTURE IN A CAST SPECIAL CR-MO STEEL HOMOGENIZED 8 HR. AT 2192°F., OIL QUENCHED AT 1652°F., DRAWN AT 1157°F. HETEROGENEOUS SORBITE IN WHICH DIFFERENCES IN CARBON CONCENTRATION APPEAR OUTLINING THE PRIMARY DENDRITIC STRUCTURE. X25.

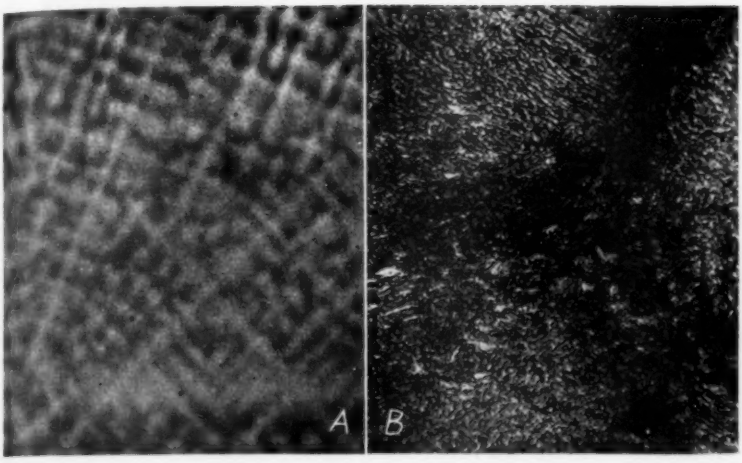

FIG. 8—(A) STRUCTURE OF A STEEL OF THE SAME TYPE AS IN FIG. 7 TREATED IN THE SAME MANNER. THE HETEROGENEOUS SORBITIC STRUCTURE HERE SHOWS THE INFLUENCE OF THE PRIMARY STRUCTURE LESS AND THE PHYSICAL PROPERTIES ARE INFERIOR. (B) SAME STEEL AS IN *A* AT A HIGHER MAGNIFICATION. THE SORBITIC STRUCTURE AND THE VARIATIONS IN CARBON CONCENTRATION ARE CLEARLY SHOWN.

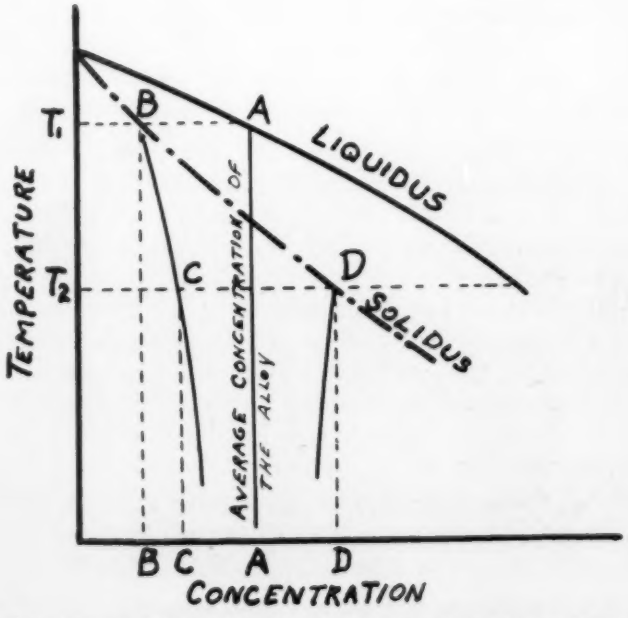

FIG. 9—SOLIDIFICATION DIAGRAM (AFTER GIOLITTI). T_1 = BEGINNING OF SOLIDIFICATION. T_2 = END OF SOLIDIFICATION.

fusion, Portevin recently proposed[5] that the structure resulting from the decomposition of the austenitic solution during the A_3-A_1 interval, which structure is modeled on that of the austenite (Fig. 3 to 6), be designated by the name of secondary or austenitic structure. He also proposed that the ferrite-pearlite structure, which is formed during the eutectoid transformation (Figs. 3 and 4) be known under the name of tertiary structure. These are the designations that we have adopted in this paper.

Primary Structure

10. The homogeneity of cast steel is evidently in close relation with the primary structure, since the cast steel always preserves its solidification structure under a form more or less attenuated. It is of interest to obtain a solidification structure which is fine or unoriented (Figs. 1 and 7). The factors affecting the fineness of the solidification structure are of a mechanical, thermal, physical, physico-chemical, or chemical[5] order.

11. Of these factors, we will consider those over which the steel foundry operator has the ability to exercise ready control, for example; thermal, chemical, and physico-chemical factors; superheating temperatures and pouring temperatures; the nature and purity of the material entering into the melting furnace, a factor on which one can not insist too much; the chemical composition of the metal; the additions of reducing elements.

12. It is known that the solidification grain of steel is refined by additions of vanadium, molybdenum, titanium (Fig. 1). A simple addition of 0.15 per cent of vanadium, for example, decidedly improves the mechanical properties of a cast steel even after a simple anneal without preceding homogenization. It is also known that nitrogen refines the grain of ferritic chrome steels. The reducing elements, such as silicon and aluminum, added as a late addition, act on the solidification grain size and even sometimes affect the phenomenon of structural heredity.

Homogenization

13. A fine solidification structure being attained, we have a metal in which heterogeneity is at a minimum in the as cast state. Nevertheless, there does exist an appreciable difference in concen-

[4]Bulletin de l'Association Technique de Fonderie, Paris, Mar., 1933.
[5]Portevin, A., "*Affinage de Structure des Alliages*," Memoire Presente au Congress International de Fonderie de Varsovie.

tration of the solid solution between the centers of crystallization or dendritic skeletons which first crystalize and the last solidified strata or interdendritic spaces. This fact, visible qualitatively to the eye (Fig. 1) or to the microscope in the primary structure (Fig. 2) or in the sorbitic structure (Figs. 7 and 8), has been clearly characterized by Giolitti in a diagram (Fig. 9)[6].

14. In this diagram the point *A*, gives the average concentration of the alloy; the point *B*, the concentration at the crystallization center at the beginning of solidification; the point *C*, the concentration of the center at the end of solidification; the point *D*, the concentration of the last layer solidified. The final concentrations at the end of solidification *C* and *D* tend to approach the average concentration due to diffusion when the latter can occur; that is, when the temperature of the metal is maintained or raised sufficiently to permit this diffusion. When the cooling of the casting is relatively rapid, the foundryman must employ a later homogenization treatment.

Table 1

INFLUENCE OF HOMOGENIZATION TEMPERATURE

	Physical Properties		
Homogenization	*Brinell Hardness* Δ	*Mesnager Resilience* Km	$\frac{\Delta}{3} + Km$
8 hr. at 2192°F.	273	8.0	115
24 hr. at 1832°F.	270	6.7	110

15. Diffusion being more rapid at high temperature, it is to the operator's advantage to carry out the homogenization treatment at high temperature. The limit, in order to avoid burning, is approximately that corresponding to the temperature at the end of solidification. In general and for high strength steels (tensile strength $>$ 128,000 lb. per sq. in.) it is not advisable to exceed 2200°F. Moreover, few foundrymen possess furnaces capable of assuring a continuous service at that temperature. In addition, however slightly, oxidizing the furnace atmosphere may be, the oxidization at high temperature causes serious damage to the castings.

16. The length of the homogenization treatment depends on the temperature. Table 1 gives the test results obtained for homogenization carried out under different temperatures and times, other factors being equal.

[6]Giolitti, F., "*Le traitment thermique preliminaire de l'acier doux et demidurs.*" Traduction francaise de Galibourg, Dunod editeur Paris. English addition published by McGraw-Hill Book Co.

17. The values in Table 1, the averages of results obtained on six different castings, prove that homogenization is much more rapid at 2192°F. than at 1832°F. The results after homogenization for 8 hr. a 2192°F. being better than those obtained after treatment at 1832°F. for 24 hr.

18. Conforming to this test we have adopted for pieces of average thickness in nickel-chrome, nickel-chrome-molybdenum and chromium-molybdenum high strength steels, containing from 0.30 to 0.40 per cent carbon, the following simple rule:

$$D = 124 - 0.1T$$

using the time D in hours at the temperature T in degrees centigrade.

19. Naturally, other rules may be considered, depending on the result desired. One French specification prescribes a homogenization annealing at a temperature of 1832°F. or above for a period of 6 to 12 hr., the time being determined by the thickness of the casting being treated.

Refinement of Secondary and Tertiary Structures

20. There is a general advantage in following the homogenization anneal by an air quench so as to derive full benefit from the refining of the tertiary structure resulting from the rapid passage through the interval $A_3 - A_1$, either by granulation of the austenite grain or by using a partial or total hardening.

Final Treatment

21. The action of the final treatment depends primarily on the chemical composition. This is a well-known fact on which we do not need to dwell except to make a few remarks of particular application to cast steels.

22. The hardening property of steel is considerably increased by certain elements, principally nickel, molybdenum and chromium. Metallurgists now can control the chemical composition of a steel destined for water, oil, or air quenching to obtain a homogeneous hardness throughout its entire mass. But the foundryman should not forget that an element favoring the penetration of hardness is capable of acting unfavorably on the fineness of the primary structure, at least in certain instances.

The foundryman then should employ combinations of alloys which, by the combination of the various elements, produce both fineness of structure and hardening, in accordance with the dimensions of the casting and the method of cooling likely to be adopted. There are still a great many points to be cleared up and there remains a vast field to be explored.

23. Another important factor relating to hardening, much discussed lately, is the "grain size." This is best determined by the method proposed by McQuaid and Ehn. The "grain size" or dimension of the grain of austenite, has only a limited and conventional significance[7]. It only represents a point of the curve $N = f(\theta et)$, in which N represents the number of austenite grains per unit of surface for a heating temperature $\theta = 1697°F.$ and a duration of heating t=8 hr.

24. In spite of possible restrictions, the knowledge of the austenitic grain size, depending always on the manufacturing method of the metal, is of undeniable interest in forged steels. In cast steels, grain size has a different significance and, according to Portevin's formula, it characterizes the aptitude to regeneration in cast steels, whereas it indicates the resistance to superheating in forged steel. On the other hand the addition of aluminum, which tends to refine the austenite grains in forged steel, on the contrary facilitates the phenomenon of "ingotism" in cast steels in which the primary structure persists[5].

HARDENING

25. Water, and even oil quenching of cast steels seems to be feared by many people. Certainly the use of self-hardening steels gives a good and safe solution. Yet this is not a reason for rejecting other means of hardening and the use of low alloy steels.

26. The problem is to establish a compromise between the kind of castings to be made, the specifications, the chemical composition of the steel, and the hardening method, in a manner to avoid internal cracks when hardening. A French specification prescribes quenching in water by immersion, or spraying, of large castings of manganese-chrome steel with a low carbon content. Oil quenching gives good results when applied to all kinds of castings of nickel-chrome and chrome-molybdenum steel which are

[7]Portevin. Metal Progress, Aug., 1937, p. 166.

not self-hardening. There is a need, naturally, in rapid cooling to provide special mountings for castings susceptible to deformation.

27. Rapid cooling has an unfavorable action on steels presenting microshrinkage and particularly on hard steels sensitive to notch effects. This action manifests itself in a lowering of the physical properties when the operations of quenching and drawing are repeated, although normally these treatments improve the quality. For example, it may be stated that, all other conditions being equal, a steel oil quenched and drawn to a hardness of 330 Brinell gives, after the first treatment, a Mesnager resilience of 5 to 6 kgm., and after a second treatment 4 to 5 kgm. only. The repetition of quenching operations ought not to be undertaken without previous checking.

FRENCH SPECIFICATIONS

28. An examination of the specifications of a country enables the formation of an opinion on the practical status of the technical problem. Table 2 gives the properties specified by the A. F. N. O. R., the French Heat Treaters Association, for cast nickel, nickel-chromium and nickel-chromium-molybdenum steels[8].

Table 2

TECHNICAL SPECIFICATIONS OF THE A. F. N. O. R.

		Elements, Per Cent			Physical Properties			
No.	*Designation*	*Ni*	*Cr*	*Mo*	*Tensile Strength, lb. per sq. in.*	*Elastic Limit, lb. per sq. in.*	*Elongation, Per Cent*	*Resilience K.UF*
I	Ni, 2 per cent	1.80	.....	.....	71,100 to 85,300	51,200	18	9
II	Ni-Cr, semi-hard	1.50	0.5	.....	102,400 to 125,100	74,000	11	5
III	Ni-Cr, Self-Hardening	3.20	1.20	0.2-0.5	128,000 to 156,500	106,700	9	6

29. These specifications are reasonable on the whole. In our opinion, they are not the maximum, particularly for class III, so we have made it a rule that the values given are an absolute minimum and we would expect a higher average. A certain dispersion of results must be expected in cast steel especially in the

[8]For nickel cast steels see: Galibourg, Centre d'Information du Nickel, Paris; Publications A-14 and A-31, Bureau of Information on Nickel, London; *"The Properties of Some Cast Nickel Steels,"* by Zima, A. G., TRANSACTIONS, American Foundrymen's Association, Vol. 41, (1933) pp. 199-223.

tensile test for hard steels. Certainly results higher than these given may be obtained, but it is well not to form an opinion on isolated tests or performances even if they may be reproduced once or twice. Cast steels will present for a long time to come a greater degree of heterogeneity than do the forged steels, but this is sometimes compensated for by the monolithic form which gives the castings a continuity advantageous in comparison with assemblies of forgings.

30. Chrome-molybdenum-vanadium and even simple chrome-molybdenum steels, contrary to the beliefs of certain authors, can, by proper handling and heat treatment, show good physical properties. This can be seen in Fig. 10 which gives a statistical report, based on 100 but determined on 200 casts, of the quality index $i = \frac{Bhn}{3} + 3 \times Mesnager\ resilience$. This equation defines correctly the specific value of a cast steel. The greatest frequency

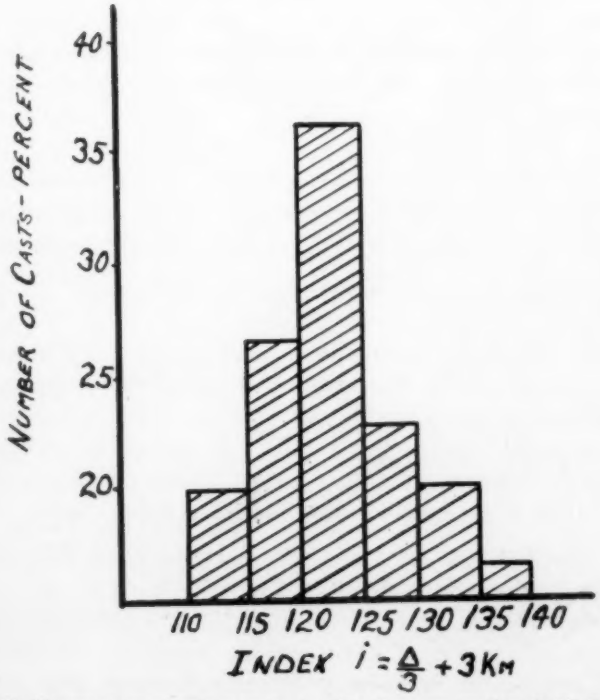

FIG. 10—VALUES OF THE INDEX $i = \frac{\Delta}{3} + 3$ KM OBTAINED IN CAST CR-MO STEELS.

for the chrome-molybdenum steel mentioned is attained with $i = 125$ to 130. The value limits equal to 110 and 140 are also high.

31. To complete the information given in this figure, it is desirable to add a table established in the same manner for the index, *tensile strength* $+ 2.5 \times$ *elongation*, which characterizes the soundness of the metal. We are only able to give (Table 3) some specifications relative to a few castings, one of good quality *B* and the other of superior quality *S*. Often in the same cast the tensile test results vary to a considerable degree, while the resilience test gives more constant values. Therefore, we have chosen to use in Table 3 only casts of steel possessing good soundness.

Table 3

PHYSICAL PROPERTIES OF CAST CR-MO STEELS

Mark	Brinell Hardness Δ	Tensile Strength, lb. per sq. in. R	Elastic Limit, lb. per sq. in. E	Elongation Per Cent A
B	300 to 310	135,100 to 142,200	116,600 to 122,300	7 to 12
S	300 to 310	139,400 to 146,500	122,300 to 128,000	8 to 14

Mark	Reduction of Area Σ	Mesnager Resilience Km	Resilience K UF	$R^* + 2.5A$	$\frac{\Delta}{3} + Km$
B	20 to 40	6 to 8	5 to 6	115 to 130	120 to 125
S	20 to 47	8 to 11	6 to 8	115 to 130	130 to 135

*R is in Kg. per mm², not in lb. per sq. in.

32. It will be noted that the index $R + 2.5A$ alone does not possess any significance when determined on coarse grain steel (annealed or superheated state), having intergranular vents which are exposed by the tensile test. Likewise, the index, $\frac{\Delta}{3} + Km$ or $R + 3Km$[9] alone has only a relative value because the resilience test does not always reveal clearly the presence of microshrinks. It is of advantage in the establishment of a specification to first make statistical tables similar to those of Fig. 10 for the two indexes previously mentioned. We need not emphasize on the elastic limit which, in the steels considered, is normally high.

33. Besides the hard and semi-hard steels of high physical properties, there is a tendency in France to use cast steels of low alloy content which have better properties than the ordinary carbon steels. Examples of these are manganese-chrome steels and molybdenum or chromium-molybdenum soft steels. The latter are

[9]Francois Le Chatelier, Revue de Metallurgie, May, 1936.

used principally as metals having a high creep limit. Their characteristics are as shown in Table 4.

Table 4

PHYSICAL PROPERTIES OF CAST STEELS HAVING LOW ALLOY CONTENT

No.	Designation	C %	Mn %	Cr %	Mo %	Tensile Strength lb. per sq. in.
1	Mo steel	0.20 to 0.25	0.8	....	0.4 to 0.5	66,900 to 71,100
2a	Cr-Mo steel	0.20 to 0.25	0.8	0.40 to 0.50	0.3 to 0.4	78,200 to 85,300
2b	Cr-Mo steel	0.20 to 0.25	0.8	0.40 to 0.50	0.3 to 0.4	78,200 to 85,300

Elastic Limit, lb. per sq. in. E	Elongation Per cent A	Resilience K UF	Creep Limit	R^*+3K	$R^*+2.5A$
49,800 to 51,200	18 to 25	9 to 11	24 to 25	78 to 82	95 to 115
52,600 to 55,500	20 to 24	10 to 12	....	95 to 105	119 to 129
52,600 to 55,500	20 to 24	6 to 7	> 24	75 to 80	109 to 110

*R is in Kg. per mm².

NOTE: No. 2a water quenched and drawn after homogenization. No. 1 and 2b air quenched and drawn after homogenization.

34. The dispersion of results in the tensile test on soft and semi-hard steels is much less than on the hard steels. We believe that this can be attributed to a lesser influence of microshrinkage; that is to say, to a lower notch effect rather than to a lesser frequency of the defect. In effect it is possible to prepare the classes II and III of Table 2 with the same nickel-chrome-molybdenum steel. Annealed to a tensile strength of 106,700 lb. per sq. in. this steel gives less dispersion than if it is treated for strength of 135,100 lb. per sq. in.

35. We conclude with this last observation, an earnest exhortation to the writers of specifications to make a distinction between cast steel and forged steel and to take into account the particular characteristics of cast steel, principally of hard cast steel, in the preparation of technical specifications.

DEFINITIONS

36. The physical properties mentioned in the test are defined below:

R=tensile strength.

E=elastic limit for a permanent deformation of 2 per cent.

A=elongation in per cent, measured on a convenient length according to the formula of the French artillery:

$$L = \sqrt{66.675}$$

S being the section of the test specimen.

Σ=reduction of area according to the formula:

$$\frac{S - s}{S} \times 100,$$ in which

S=initial section of the test specimen.

s=section of the constricted part of the fracture.

Km=Mesnager resilience of a specimen 1.0 x 1.0 x 5.5 cm. (0.3937 x 0.3937 x 2.1654 in.) notched 2 mm. (0.07874 in.) deep, finished with a radius of 2 mm. (0.03937 in.)

$K\ UF$=resilience according to the Unified French specification, on a specimen 1.0 x 1.0 x 5.5 cm. (0.3937 x 0.3937 x 2.1654 in.) with a notch 5 mm. (0.19685 in.) deep, finished by a radius of 2 mm. (0.07874 in.)

$\triangle$=Brinell hardness, 3000 Kgs. load on a ball of 10 mm. diameter.

F_1=creep limit, 0.0015 permanent deformation between the 25th and 35th hr. at 400°C. (752°F.)

Recommended Practice for Sand Cast High Lead Bronzes

Preface

This is the fourth in a series of recommended practices being sponsored by the Non-Ferrous Division of the American Foundrymen's Association, the first being issued in 1934 (Transactions A.F.A., vol. 42, pp. 1-26), the second in 1936 (Transactions A.F.A., vol. 44, pp. 33-50) and the third in 1938 as (A.F.A. preprint No. 38-4). Others covering the various groups of alloys listed in the Cast Metals Handbook (A.F.A. 1935), Non-Ferrous Section, will be prepared and published in the future.

Although the various practices which are treated are recommended by those considered specialists in the various fields, these practices are not intended as specifications. The alloy numbers have no particular significance other than to differentiate alloys in this particular specification. Wherever possible, those numbers are identified with existing alloy specifications.

The publishing of data relative to various alloys and treatment of such alloys by the American Foundrymen's Association, does not insure anyone using such data against liability for infringement of any patents that may now exist or are pending. It should also be understood that the publication of data concerning patented alloys or processes does not constitute a recommendation of any patent or proprietory rights that are involved.

In preparing these recommendations, the Division has attempted to collect, through reliable sources, such information as will be of practical use to foundrymen handling non-ferrous alloys.

This report is limited to sand cast high lead bronzes and is divided as follows:

A. General Recommendations

1. *Molding—(a) Sand, (b) Facing, (c) Gating and Heading (d) Cores.*
2. *Melting and Pouring—(a) Types of Melting Equipment, (b) Precautions, (c) Fluxing and Deoxidizing, (d) Melting and Pouring Range.*

NOTE: This Recommended Practice was originally presented at the annual business meeting of the Non-Ferrous Division at the Cleveland Convention, May 16, 1938 and was released for Publication as a Recommended Practice, following the meeting of the Non-Ferrous Division Recommended Practices Committee at The 1939 Cincinnati Convention, May 15, 1939.

3. *Finishing.*
4. *Heat Treatment.*
5. *Defects, Their Cause and Prevention—(a) Melting Defects, (b) Pouring Defects, (c) Molding Sand Defects, (d) Dumping Defects, (e) Cleaning Defects, (f) Defects Caused by Impurities in the Metal.*

B. *Properties and Applications*
1. *Chemical Control Limits.*
2. *Typical Properties of Sand Castings.*
3. *Conforming Specifications.*
4. *Development and Field of Use.*

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A. GENERAL RECOMMENDATIONS

1. *Molding*

(a) *Sands.* The high lead alloys, particularly the alloy 80 per cent copper, 10 per cent tin, 10 per cent lead (No. A-42), with phosphorus added, possess the property of penetrating deeply into the sand unless materials are used as facings to prevent it. Satisfactory sands for these alloys are those falling within the A.F.A. classification 2E to 1F. Coarser sands are not recommended unless a suitable facing is used to prevent metal penetration.

The above sands should provide the following physical characteristics:

Compressive Strength, lb. per sq. in.	4- 8
Permeability (A.F.A.)	5- 25
Clay Content—per cent	10- 20
Fineness Number (A.F.A.)	150-230

Sands providing the above characteristics will be found to give a smooth finish to the castings and are open enough to allow the escape of steam and gases from the mold when making small castings. However, it will be found necessary to vent molds freely for heavier castings.

For extremely heavy castings or castings of intricate design, it is sometimes found advantageous to make dry sand molds.

(b) *Facing.* The facing recommended for high lead alloys is the best grade plumbago available. Cheap grades of plumbago are likely to cause more harm than good.

(c) *Gating and Heading* (See Fig. 1). Gating of the high lead alloys is of the simplest brass or bronze variety. For heavy castings, it is found advantageous to make the sprue and gate in core sand. This is to prevent washing of sand from sprue or gate into the mold cavity. Skimmer cores also may be employed to hold back dross and prevent its entry into the mold. Gates should be

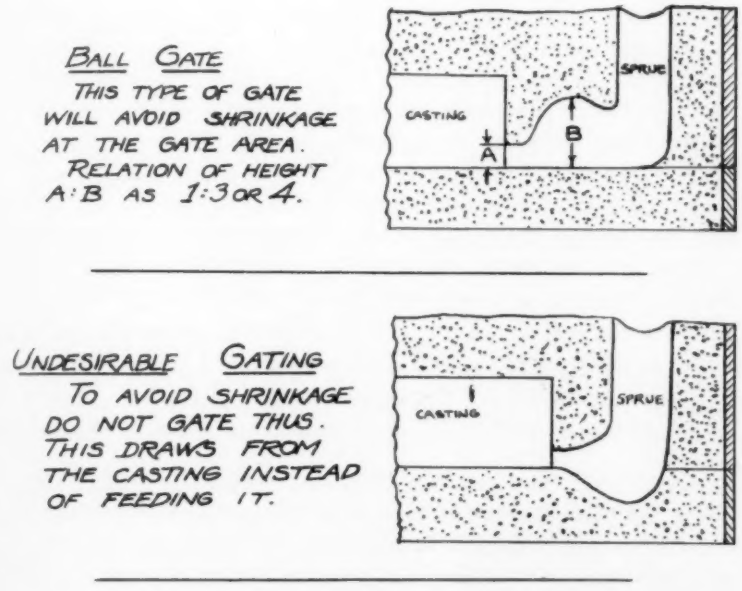
BALL GATE
THIS TYPE OF GATE WILL AVOID SHRINKAGE AT THE GATE AREA.
RELATION OF HEIGHT A:B AS 1:3 OR 4.
SPRUE
CASTING
A
B
UNDESIRABLE GATING
TO AVOID SHRINKAGE DO NOT GATE THUS. THIS DRAWS FROM THE CASTING INSTEAD OF FEEDING IT.
CASTING
SPRUE
BLIND RISER
TO PREVENT SHRINKAGE
THE AREA OF THE UPPER PART OF THE RISER MUST BE LARGER THAN CONTACT AREA AT CASTING. VENT MUST BE LARGE ENOUGH TO RELEASE ENTRAPPED GAS.
VENT
ARCH IS STRONGER THAN FLAT
BLIND RISER

FIG. 1 — METHODS OF GATING HIGH LEAD BRONZE CASTINGS

cut so that they are smaller than the sprue to allow the sprue to be kept full during pouring.

High lead alloys have a long range of solidification due to their complex composition and are not entirely solid until the freezing point of lead is reached (621°F.). For this reason, apparent shrinkage of the casting is slight and what shrinkage there is, is distributed over a large area as an inter-crystalline shrinkage rather than as a single void in the casting.

This happy property is taken advantage of by most high lead bronze founders and only the heavier castings are found to need risers. Castings so made, make excellent bearings as the openness of the metal creates cavities which greatly assist in maintaining an oil film on the bearing surface. However, these castings are made at a sacrifice of physical properties and, should maximum physical properties be required in the casting, ample provisions for proper risers and gates should be made and in general a lower pouring temperature should be used. Gates and risers should be adjusted so that the casting will solidify at a uniform rate in thin and heavy sections. Risers should have sufficient volume to remain liquid while the casting is solidifying. If cavities appear in the casting after the removal of the risers, then either the risers were too small or the metal in them was too cold and solidified before the casting. A small amount of ¼-in., graded charcoal placed on the top of the riser helps to prevent the air from cooling the riser too rapidly.

High lead alloys are hot short and frequently cracks will occur in the cold castings at the intersection of light and heavy sections. This condition can be overcome by increasing the fillets at such junctions or by placing a riser on the heavy section and gating into the light section.

In general, it may be said that light castings do not require risers while heavy castings need risers to prevent shrinkage cracks and depressions on the surface. This will usually apply to bearings. Other castings, particularly pressure castings, require more care in the placing of gates and risers to insure maximum soundness.

(d) *Cores.* All cores should be well vented and covered with plumbago to prevent metal penetration. Any typical general purpose core mixture providing the desired permeability and strength may be used.

2. *Melting and Pouring*

(a) *Types of Melting Equipment.* All forms of melting equipment, including oil, gas or coke fired crucible furnaces; electric furnaces; oil or gas-fired, open-flame, stationary or rotary furnaces, and even cupola furnaces, can be successfully used to melt the high lead bronze alloys.

(b) *Precautions.* Certain types of furnaces are most readily adapted to the production of metal free from absorbed gases and oxidation than others. Furnaces of this type are those in which the metal does not come in contact with the fuel or the products of combustion, such as in the crucible type of furnace. By placing a cover on the crucible or a flux on the metal, or both, little opportunity is given the metal to absorb gas or become oxidized.

The color of the flame in an open flame furnace is the best guide to the condition being maintained in the melting furnace. For the most satisfactory results, the furnace atmosphere should be maintained as nearly neutral as possible. A green flame denotes an oxidizing atmosphere which will result in excessive oxidation of the metal. A flame that is too yellow or smoky indicates a reducing atmosphere and one which may result in gas absorption by the metal.

Both solid and gaseous fuels are sometimes high in sulphur and, to prevent contamination from this element, it may be found necessary to employ a slag covering for the metal. Ordinary bottle glass will serve very well.

Refractories should be of high quality and resistant to the corrosive action of the metals and their oxides.

Preheating of air or fuel, or both, will give faster melting and less time and opportunity for the metal to become contaminated during melting.

(c) *Fluxing and Deoxidizing.* Soda ash, borax, glass, charcoal, phosphor tin and phosphor copper are the fluxes most commonly recommended for use with high lead bronzes. Soda ash and borax are used to collect and make for more fluid slags. Since too fluid slags are difficult to skim, these materials should be used carefully. Sand or glass is often added as a means of controlling the slag fluidity and at the same time appears to reduce the attack of the soda ash and borax on crucibles and furnace lining.

Although charcoal is so often recommended as a protective flux during melting, there is always a danger of introducing gas into the metal, particularly if the charcoal is moist or only partly burned. A slag composed of soda ash, borax and sand or glass is therefore recommended, the mixture to be that which best suits the conditions of the individual foundry.

Phosphor-tin and principally phosphor-copper are used to deoxidize these bronzes. Metal deoxidized with phosphor-copper will be found to be more fluid with the formation of little dross during pouring. It can be added in the ladle immediately before pouring or in the furnace charge, whichever experience shows to be best.

Zinc is sometimes added to alloy A-42 to increase the fluidity and has moderate deoxidizing powers. The amount of zinc added is limited by the specification, in this case a maximum of 0.5 per cent being permitted.

(d) *Melting and Pouring Range.* In making these alloys from virgin metal, the copper should be melted first and then the lead added, followed by the tin and zinc and finally phosphorus as phosphor-copper. If the phosphorus content is high, some is added after the lead and the remainder just before pouring. If mixed scrap is used, heavy pieces are melted first and so form a bath for the smaller pieces or turnings.

Melting of these high lead bronze alloys actually starts at 621°F., the melting point of lead. No definite pouring temperature can be given for any one alloy as this will depend on the nature of the mold and the size of the casting. The lowest practical pouring temperature is about 1750°F., although the average is about 2000°F. The following table indicates the permissible variations in the pouring temperatures of the various alloys under consideration.

Alloy		Pouring Range
A 41		1900 to 2300°F.
A 42		
	Phosphorus	
	0.0 — 0.02 per cent	1900 to 2200°F.
	0.10 — 0.20 per cent	1840 to 1900°F.
	0.50 — 1.00 per cent	1750 to 1840°F.
A 43		1850 to 2200°F.
A 44		1850 to 2200°F.

Temperature measurements are best made with a pyrometer with a suitable scale calibration to cover the range of temperatures employed. Such an instrument may be equipped with either an open or closed end thermocouple, the latter either with or without a protective tip, depending on the preference of the operator. When using an unprotected-tip couple, care must be exercised to prevent bridging of the wires with a metal or oxide film. By immersing the wires to a depth of 4 to 6 in. until they have reached the metal temperature and then withdrawing them quickly until the tips remain just under the surface, reading the instrument at this time, will help to eliminate errors due to this condition. For this reason, modern protected tip couples are finding increasing favor for use with brass and bronze.

3. *Finishing*

High lead bronze castings may be trimmed in much the same way as other non-ferrous castings. Gates and risers are preferably removed with a metal band saw. Breaking these off while the metal is still in the hot-short state is also practiced, although this procedure requires considerable care to produce a fracture that is not too close to the casting. When gates and risers are inaccessible or are too large for sawing, they can be readily removed by chipping.

Cleaning may be accomplished by tumbling and by blasting. The latter method is particularly satisfactory, using either sand or steel grit as the blasting material.

Such final cleaning as may be necessary can be done with the usual chisels, files, and grinding and polishing wheels provided for this purpose.

4. *Heat Treatment*

Heat treatment has little effect on the physical or mechanical characteristics of these alloys and, therefore, is not used commercially.

5. *Defects, Their Cause and Prevention*

(a) Melting Defects

1. *Gas inclusions* in castings may be due to the absorption of reducing gases during melting and which are released on solidifica-

tion of the metal. Such inclusions are revealed as bright, round holes on machined or fractured faces.

Prevention—Maintain as nearly a neutral atmosphere as possible in the furnace during melting. Fuels high in sulphur sometimes cause gas holes.

2. *Dross Inclusions,* which may be due to oxidized metal, are revealed on machined or chipped surfaces as greenish colored areas or in fractures as red to dark brown inclusions.

Prevention—Melt under a neutral atmosphere or, if metal is known to have been oxidized, flux with phosphor-copper, the amount depending on the degree of oxidation. Molten oxidized metal is easily distinguished by its tendency to dross heavily on the surface. It appears colder than unoxidized metal when at the same temperature and pours with less fluidity and considerable dross forming on the sprue head. Dendritic fractures are sometimes caused by oxidized metal. A ring of holes at about 3/16-in. depth found against a dried sand surface has been known to have been caused by oxidized metal rather than by excess moisture. Discolored fractures are frequently due to the intercrystalline shrinkage which permits the access of oxidizing gases, thus turning the crystal surfaces various colors.

(b) Pouring Defects

1. *Misrun* or *mold cavity incompletely filled* denotes too low a pouring temperature for the design of casting and its gating methods.

Prevention—Pour at a higher temperature.

2. *Cold shut* or *apparently cracked* castings which on fracture reveal an area or line along which the metal has failed to join, can be caused by too low a pouring temperature, pouring too slowly or a stop in the pouring.

Prevention is indicated by the causes.

3. *Sand wash, fins, excessive burning in* and castings generally hard to clean may be defects caused by pouring at too high a temperature.

Prevention—Pour at the best determined temperature.

4. *Dross* and *dirt inclusions* may be caused by careless pouring, skimming, or lack of deoxidation as well as other causes. (See also dross inclusions under Melting Defects.)

Prevention—Pour steadily, skim thoroughly and pour as close to the sprue head as possible.

(c) Molding Sand Defects

1. *Shift* caused by cope and drag flasks not matching.

Prevention—Keep flasks and pins in good condition.

2. *Crush* caused by a drop of an area of the mold surface into the mold cavity on closing mold or in placing of weights on the mold.

Prevention—Close molds more carefully and investigate sand conditions and condition of bottom boards and flasks.

3. *Variation in wall thickness* may be caused by shift of flasks or careless molding.

Prevention—More careful molding.

4. *Sand wash* is a result of soft ramming, weak or wet sand.

Prevention—Ram harder, use a facing and investigate sand conditions.

5. *Sand blow* may be caused by wet sand, sand too close, or improperly vented mold.

Prevention—Maintain sand at proper moisture, use a more open sand or vent more freely.

6. *Core blows* are generally caused by damp cores or badly vented cores.

Prevention—Dry cores thoroughly and vent more freely.

7. *Scabs* result from uneven ramming or loose sand in mold.

Prevention—Ram evenly and thoroughly clean mold.

8. *Burning into sand* is often caused by soft ramming or sand too coarse and open.

Prevention—Suitable ramming, use of correct grade of sand, or both.

9. *Burning into cores* is caused by wrong core mixture or core too open.

Prevention—Use correct core mixture and blacken cores.

10. *Shrinkage cracks or cavities* may be caused by misplacing of gates and risers or the use of risers that are too small.

Prevention—Study nature of castings as to wall thickness and filletting and then decide the best possible manner of gating and heading.

(d) Dumping Defects

1. *Lead sweat* is caused occasionally by removing castings from the sand at too high a temperature. Lead sweat of this nature will not usually cause serious casting defects.

Prevention—Allow the casting to cool to a lower temperature before removing from the sand. Lead is liquid in the casting until it reaches a temperature below 621°F.

2. *Cracked or broken castings* are often caused by careless handling when castings are removed from the sand while in the hot short condition.

Prevention—Exercise more care when removing castings from sand and allow to cool longer in the mold before removal.

(e) Cleaning Defects

All cleaning defects are caused by careless handling of the cleaning equipment, fracturing gates and risers into the casting itself or from insufficient knowledge of the size and shape of the casting.

(f) Defects Caused by Impurities in Metal

1. *Phosphorus,* in addition to being beneficial as a deoxidizer and densifier of bronze, may be a harmful impurity unless its presence is known. It causes excessive shrinkage, burning into sand, and fusion of the sand to the surface of the castings making them difficult to clean and machine. As low as 0.03 per cent phosphorus in the alloy A-43, will cause excessive burning into sand and excessive shrinkage when the castings are poured at normal temperatures.

2. *Silicon* may be classed as the most harmful of impurities in high lead bronze. Its presence may sometimes be detected easily

in the molten metal due to dark patches of dross continually rising to the surface. When a casting is being poured, a swirling of light white material occurs in the sprue. When the castings are removed from the sand, they have the appearance of having been whitewashed. The surface will be very rough, coated with a white silicious material and accompanied by much lead sweat. The fracture of a leaded bronze containing silicon is golden to red brown in color, dendritic and usually with a separation of the metal along the length of the casting. The openness of the metal allows the lead to sweat freely. Even traces of silicon should be avoided.

3. *Aluminum* is almost as harmful as silicon. Its presence may be detected in the molten metal by the strength of the oxide film which forms on the surface of the metal. The resultant castings will be silvery white in appearance and appear to be wormeaten with the characteristic dross of aluminum bronzes. Aluminum causes a similar structure to that of silicon which will be dendritic, causing the lead to sweat from the casting. Even traces of aluminum should be avoided.

4. *Antimony* causes no outward defects but tends to a dendritic fracture and should not be allowed in bronze in amounts in excess of 0.75 per cent. It has the property of increasing the resistance to pounding of the bronze but its resistance to impact is decreased. If added in excessive amounts, antimony will make the bronze brittle. It is better to keep this element as low as possible.

5. *Iron,* if present in large amounts, causes hard spots, thus making the castings difficult to machine. It may be found both in solution or in the free state and solution. It is known to cause excessive internal shrinkage and discolors the alloy fracture yellow red to brown.

6. *Sulphur* should be kept to a minimum as it causes pinholed or gassed castings. Fuels are the usual source of contamination with this element. Contamination should be guarded against by the use of a flux or covering for the metal during melting. It is best to specify a low sulphur content when purchasing fuel for melting purposes. The determination of sulphur in bronze is a very simple matter, the usual evolution method as for steel giving sufficiently accurate results.

B. Properties and Applications

1. *Chemical Control Limits*[1]

Table 1

CHEMICAL CONTROL LIMITS FOR SAND CAST HIGH LEAD BRONZE CASTINGS

A.F.A. Alloy No.		A 41	A 42	A 43	A 44
Copper—per cent	Minimum	84	79	72	70
	Desired	85	80	73	70.5
Tin—per cent	Minimum	4.25	9	7	5.25
	Desired	5.00	10	8	6
	Maximum	5.75	11	9	6.75
Lead—per cent	Minimum	8	9	13.5	18
	Desired	9	10	15	20
	Maximum	10	11	16.5	22
Zinc—per cent	Minimum	0.5	....	1	0.5
	Desired	1	....	1.5	1
	Maximum	2	0.50	2	1.5
Phos.—per cent	Minimum	....	0.05	....	....
	Desired	....	....	....	....
	Maximum	0.01	1.00	0.05	0.05
Antimony—per cent Maximum		0.25	0.50	1.00	1.00
Iron—per cent	Maximum	0.25	0.25	0.25	0.25
Nickel—per cent	Maximum	0.50	0.25	0.50	0.50
Sulphur—per cent Maximum		0.08	0.08	0.08	0.08
Aluminum—per cent Maximum		0.005	0.005	0.005	0.005
Silicon—per cent Maximum		0.03	0.03	0.03	0.03
Other than copper, tin, lead, zinc, antimony, and nickel —per cent		0.50	0.50	0.50	0.50

[1]These compositions apply to the alloys in ingot form.

The mechanical properties listed in Table 2 are based on machined specimens cut from the test castings. The castings were made in dry sand molds without facing or blackening. The pattern was equipped with permanent gates to eliminate variations due to hand cutting. The pouring temperature in the case of alloys A 41, A 43, and A 44 was 1900°F. The pouring temperature of test

2. *Typical Physical Properties of Sand Castings*[1]

Table 2

TYPICAL PHYSICAL PROPERTIES OF SAND CAST HIGH LEAD BRONZES

A.F.A. Alloy No.	A 41	A 42	A 43	A 44
Ultimate Tensile Strength,[2] lb. per sq. in.	25,000 to 34,000	27,000 to 33,000	25,000 to 30,000	22,000 to 27,000
Proportional Limit in Tension,[2] lb. per sq. in.	7,000 to 9,000	9,500 to 11,500	6,000 to 9,000	5,000 to 8,000
Yield Strength in Tension,[2] lb. per sq. in.	14,000 to 15,000	18,500 to 20,000	14,000 to 17,000	13,000 to 16,000
Elongation in 2 in.[2]—per cent	15 to 25	7 to 12	11 to 20	10 to 16
Reduction in Area[2]—per cent	12 to 20	8 to 13	11 to 15	9 to 14
Compression of 0.001 in.,[3] lb. per sq. in.	10,500 to 12,000	12,500 to 15,000	12,000 to 15,000	11,000 to 14,000
Compression of 0.1 in.,[3] lb. per sq. in.	29,000 to 35,000	45,000 to 47,000	38,000 to 40,000	32,000 to 35,000
Compression at 100,000 lb.[3] per. sq. in.—in.	0.34 to 0.41	0.27 to 0.30	0.30 to 0.33	0.35 to 0.40
Modulus of Elasticity, lb. per sq. in.	.....	8,500,000	8,800,000	.....
Brinell Hardness No. (500 Kg. load—100 mm. ball)	45 to 60	50 to 70	45 to 65	40 to 55
Patternmaker's Shrinkage, in. per ft.	$\frac{3}{16}$	$\frac{3}{16}$	$\frac{3}{16}$	$\frac{3}{16}$
Specific Gravity	8.7	8.9	9.4	9.6
Weight—lb. per cu. in.	0.314	0.322	0.335	0.337
Izod Impact Resistance, ft. lb.[4]	.....	2 to 4	4 to 5	.....

[1] Mechanical properties based on specimens machined from test castings and not to be used for specification.

[2] Tensile specimens machined to provide standard ½-in. diameter by 2-in. gage section. (See Fig. 2).

[3] Compression specimens machined to 1.128-in. diameter by 1-in. high.

[4] Impact specimens machined to standard Izod cantilever dimensions, 10 mm. x 10 mm. with 45° V-notch.

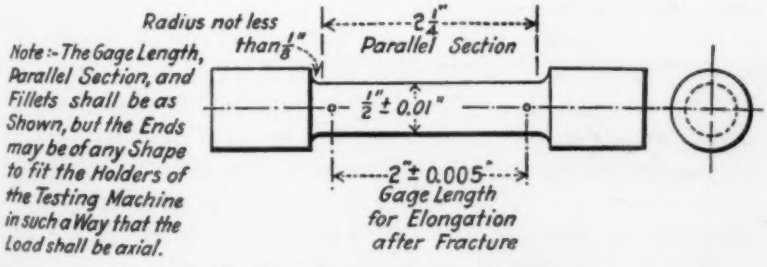
Radius not less than 1"/8
2 1/4" Parallel Section
Note:- The Gage Length, Parallel Section, and Fillets shall be as Shown, but the Ends may be of any Shape to fit the Holders of the Testing Machine in such a Way that the Load shall be axial.
1/2" ± 0.01"
2" ± 0.005"
Gage Length for Elongation after Fracture

FIG. 2 — DIMENSIONS OF MACHINED SPECIMENS

blocks in alloy A 42 varied with the phosphorus content, from 1900°F. in the case of a melt where phosphorus was added only as a deoxidizer and therefore present in the minimum amount permissable, to 1750°F. with 1 per cent phosphorus present.

3. *Conforming Specifications*

Table 3

CONFORMING SPECIFICATIONS FOR SAND CAST HIGH LEAD BRONZE

Alloy A 41—American Society for Testing Materials No. B-30-36, Grade 15.
American Society of Automotive Engineers, Alloy No. 66.

Alloy A 42—American Society for Testing Materials No. B30-36, Grade 16.
American Society of Automotive Engineers, Alloy No. 64.

Alloy A 43—American Society for Testing Materials No. B30-36, Grade 18.

Alloy A 44—American Society for Testing Materials No. B30-36, Grade 20.

4. *Field of Use*

The high lead alloys provide excellent anti-frictional qualities. As a result of these superior properties, they have replaced the pure tin bronzes for many bearing applications where lubrication is not of the first order and climatic conditions are severe, such as in railroad bearings. The high lead alloys also are used successfully for acid resisting castings in the mining and paper industries.

Alloy A 41 is used principally for small bearings, such as automotive bearings and bushings, particularly those to be machined by broaching. Alloy A 42 is used for heavy pressure castings which are to be finished by methods other than broaching. Alloy A 43 is used as a general utility bearing bronze, for acid-resisting castings and locomotive wearing parts. Alloy A 44 is used principally for journal bearings and bearings operated at high speeds and moderate pressures.

Atmosphere Gases for Malleable Iron Annealing and Their Application

By John Dow,* Detroit, Mich.

Abstract

The manufacture and general properties of the more commonly used atmosphere gases are discussed briefly. Tests are described and results given of the action of carburizing and decarburizing atmospheres on machined and unmachined surfaces of iron. Results on machines' faces indicate a carburizing gas slightly increases the graphitizing rate of the layer of iron affected by the gas, and that a decarburizing gas markedly reduces the rate of graphitizing. Results on unmachined surfaces indicate a slower graphitizing rate in the sub-surface layer than in the core iron irrespective of the type of gas used. The application of atmosphere gases to production annealing of both regular and pearlitic malleable is discussed. Furnace design, gas inlet location, effect of furnace leaks, and the principles involved are given consideration.

Introduction

1. When annealing malleable iron without pots, it is necessary to surround the castings with an atmosphere that will not scale the iron. There are several types of gas generators on the market, each capable of transforming the available fuel gas into a low B. t. u. gas, non-scaling to iron. The gases from the various types of generator have certain individual characteristics, and differ, from each other, mainly, in their action on the carbon in the iron. Gases may be actively carburizing, actively decarburizing, or practically neutral to the carbon in the iron.

2. Most of the production furnaces now in operation produce castings having evidence of both decarburization, and untransformed pearlite in a layer of varying thickness just under the surface of the casting. There may be from 0.010 to 0.015 in. of complete decarburization backed by another layer of similar thickness containing varying amounts of pearlite, lying outside of a completely annealed core.

3. The purpose of this paper is to discuss the analysis and properties of the more commonly used atmospheric gases; to

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report the results of some tests showing the effect of these gases on the rate of annealing of the sub-surface layer of the iron; and to discuss the manner in which each type of gas may best be applied to production furnaces.

CLASSIFICATION OF ATMOSPHERE GASES

4. All gases used as an atmosphere in malleable iron annealing are either carburizing or decarburizing in their action on the iron, or are to all practical purposes neutral in this regard. Methods of manufacturing these different types of gases and sample analyses will be given under separate headings.

Decarburizing Gases

5. The type of atmosphere gas machine most commonly used in malleable iron annealing is the simple, well-known, bright-annealing, gas generator, consisting of a gas-air mixer, cracking chamber, and water condenser. A diagrammatic sketch of this machine appears in Fig. 1. The analysis of the gas made in this machine can be varied to some extent. But, as long as the gas produced is maintained non-scaling in its action, it will actively decarburize the iron being annealed, except during the last few hours of the annealing cycle, when the temperature is low and there is very little combined carbon left for the gas to act on.

ANALYSIS OF GAS PRODUCED BY A SIMPLE GENERATOR WITH LIMITING MIX RATIOS

Table 1

Mix Ratio								
Natural Gas;		*Air*	H_2O	CO_2	CO	H_2	CH_4	N_2
1	:	4	3.0	4.0	12.5	15.1	4.0	61.4
1	:	7	3.0	5.3	10.0	14.0	0.6	67.1

6. The analysis of the gas produced by this machine is varied by altering the gas-air ratio fed to it. The limit of operation on the rich side is determined by the fact that the gas-air mixture must be combustible, and on the lean side by the gas formed becoming oxidizing to iron. The limiting mix ratios and analyses produced are approximately as shown in Table 1.

7. Some generators on production furnaces are being operated with mix ratios that produce gases oxidizing in character, with apparently better results than when richer mixtures were

used. This method of operation will be discussed later in the paper. It is sufficient at present to make the observation that the analysis of the gas produced by the generator is altered by reaction with the carbon of the iron in the furnace, so as to become non-scaling in character.

Neutral Gases

8. To the best of the writer's knowledge no production furnaces are as yet being supplied with a gas which may be con-

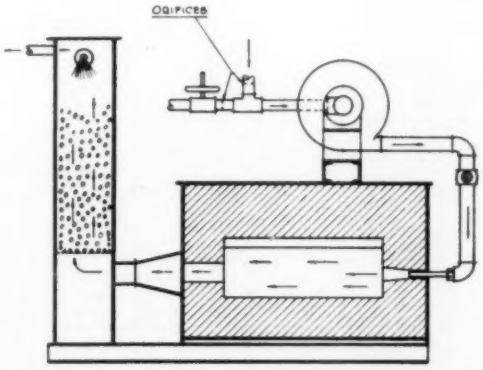

FIG. 1—SKETCH OF A SIMPLE, BRIGHT-ANNEALING, GAS GENERATOR, CONSISTING OF A GAS-AIR MIXER, CRACKING CHAMBER, AND WATER CONDENSER.

sidered to be practically neutral to the carbon in the iron being annealed.

9. Before defining the term "practically neutral" it is necessary to consider the possible carburizing and decarburizing rates at the temperatures of the first and second stages of the annealing cycle. In 1 hr. at 1750°F., a common first stage annealing temperature, it is possible to carburize or decarburize as deep as is possible in approximately 24 hr. at 1350°F., an approximate average of the second stage temperature range. For this reason it is necessary for practical purposes to consider the effect of the gas on the iron only during the first stage of the annealing cycle. Furthermore, since the combined carbon in the iron remains high during this first stage, it is possible to produce gases which will have little or no carburizing or decarburizing action during this part of the cycle. These gases may be termed "practically neutral."

10. In industry, atmospheres, which may be termed neutral to high carbon steel or iron, are being used mainly in furnaces for clean hardening high carbon or previously carburized steel parts such as gears, piston pins, etc. No decarburization is permissible during the hardening. This type of atmosphere gas is being manufactured by two methods.

11. The first of these machines built for a large continuous furnace was installed at the Buick Motor Company in 1935 and is

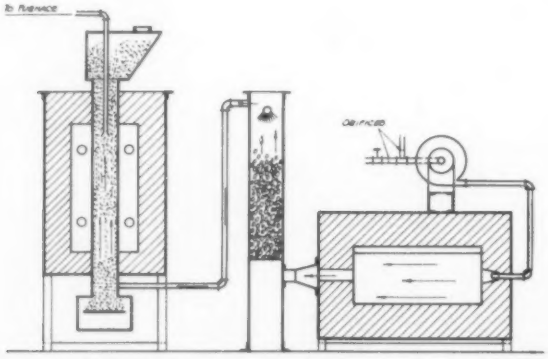

FIG. 2—SKETCH OF A CHARCOAL GAS GENERATOR CONSISTING OF A GAS-AIR MIXER, CRACKING CHAMBER, WATER CONDENSER, AND AN EXTERNALLY HEATED RETORT FILLED WITH CHARCOAL.

popularly known as a "charcoal gas generator." A diagrammatic sketch of such a generator appears in Fig. 2. The generator consists of an externally heated retort filled with charcoal, through which is passed a gas of the type produced by the bright-annealing gas generator previously described.

12. In its passage through the hot bed of charcoal, the bright-annealing gas is allowed to complete its tendency to react with carbon. In so doing, the gas undergoes a change in composition and becomes one having the following approximate analysis: H_2O 0.2 per cent, CO_2 0.2 per cent, CO 20.0 per cent, H_2 22.0 per cent, CH_4 2.0 per cent, N_2 55.6 per cent. This gas differs from the bright-annealing gas only in that the two decarburizing gases — water vapor and carbon dioxide — have been converted almost entirely to carbon monoxide and hydrogen. This gas may be considered neutral in its affect on the carbon of the iron as a result.

13. Another method of producing a gas somewhat similar in analysis and action, is to treat bright-annealing gas chemically to

remove the two decarburizing constituent gases — carbon dioxide and water vapor. An Ethanol-amine solution is used to absorb dioxide and activated alumina is the adsorbent usually used for water vapor.

Carburizing Gases

14. Because of limitations imposed by the volume of gas usually required by malleable annealing furnaces there are only two methods of manufacturing a carburizing gas which need be considered. These two methods are alike in that they embody the addition of a hydrocarbon such as natural gas or propane to one of the neutral gases listed above or to bright-annealing gas.

15. If the hydrocarbon, bright-annealing gas mixture be used, it is necessary to feed the mixture to the hot zone of the furnace. Here the two gases react and the decarburizing agents disappear. Because the necessary amount of hydrocarbon addition is large, a rather heavy deposition of carbon occurs both on the work and heating tubes or elements in the furnace. This deposition of carbon can become very objectionable. Only 1 to 5 per cent of hydrocarbon added to charcoal gas is sufficient to give this gas a noticeable carburizing action. Very little soot deposit occurs in this instance.

ANNEALING TESTS

16. In order to determine the effect of the above gases on malleable iron being annealed, samples of white iron were held in first, a decarburizing atmosphere, and second, a carburizing atmosphere. The samples were held at first stage annealing temperatures for a longer time than is normal for the iron used. The bars were then removed and allowed to cool. However, before subjecting the test bars to this treatment, ⅛ in. was machined off one side of each white iron bar. After removing micro-samples from each bar, they were then cut in two and the four resulting test bars treated as follows: One piece of each original bar was given a protective coating of non-case and packed in a small pot of spent carburizing compound. These samples were then given an accelerated second stage annealing, consisting of heating to 1600°F., then cooling and holding at 1400°F. for 3 hr., followed by further cooling and holding at 1325°F. for 2 hr. An incompletely annealed structure was desired in order that the progress

of the annealing in the core and surface layers could be compared, and relative annealing rates determined.

17. The second piece of each bar was given a similar time and temperature treatment in a gas similar in character to that to which it had been exposed during the prolonged first stage. A tabulation of the tests would be as shown in Table 2.

18. The first stage annealing in a decarburizing atmosphere was done in a production malleable annealing furnace in which the atmosphere analyzed approximately 6 per cent CO_2 and may be considered fairly actively decarburizing. At the end of 24 hr. at 1750°F. a micro-sample showed total decarburization to a depth of 0.012 in. under the rough cast surface and 0.016 in. under the machined face. Back of the total decarburization was a layer which etched darker being apparently higher in combined carbon. This layer varied in depth on the cast faces and the average was somewhat lighter than on the machined face.

19. Samples 1 and 2 were almost identical both showing total decarburization of 0.017 to 0.018 in. on all surfaces. This was backed by a layer high in pearlite, 0.030 to 0.035 in. deep on the machined face and 0.055 in. deep on the as-cast faces. A micrograph typical of the condition on the cast and machined faces is shown in Fig. 3.

Table 2

TABULATION OF ANNEALING TESTS

	First Stage						
Sample No.		*Time, Hr.*		*Temp., °F.*			*Remarks*
1		24		1750			decarburizing gas
2		24		1750			decarburizing gas
3		48		1680			carburizing gas
4		48		1680			carburizing gas
	Second Stage						
Sample No.	*Temp., °F.*	*Time, Min.*	*Temp., °F.*	*Time, Hr.*	*Temp., °F.*	*Time, Hr.*	*Remarks*
1	1600	10	1400	3	1325	2	decarburizing gas
2	1600	10	1400	3	1325	2	surface protected
3	1600	10	1400	3	1325	2	carburizing gas
4	1600	10	1400	3	1325	2	surface protected

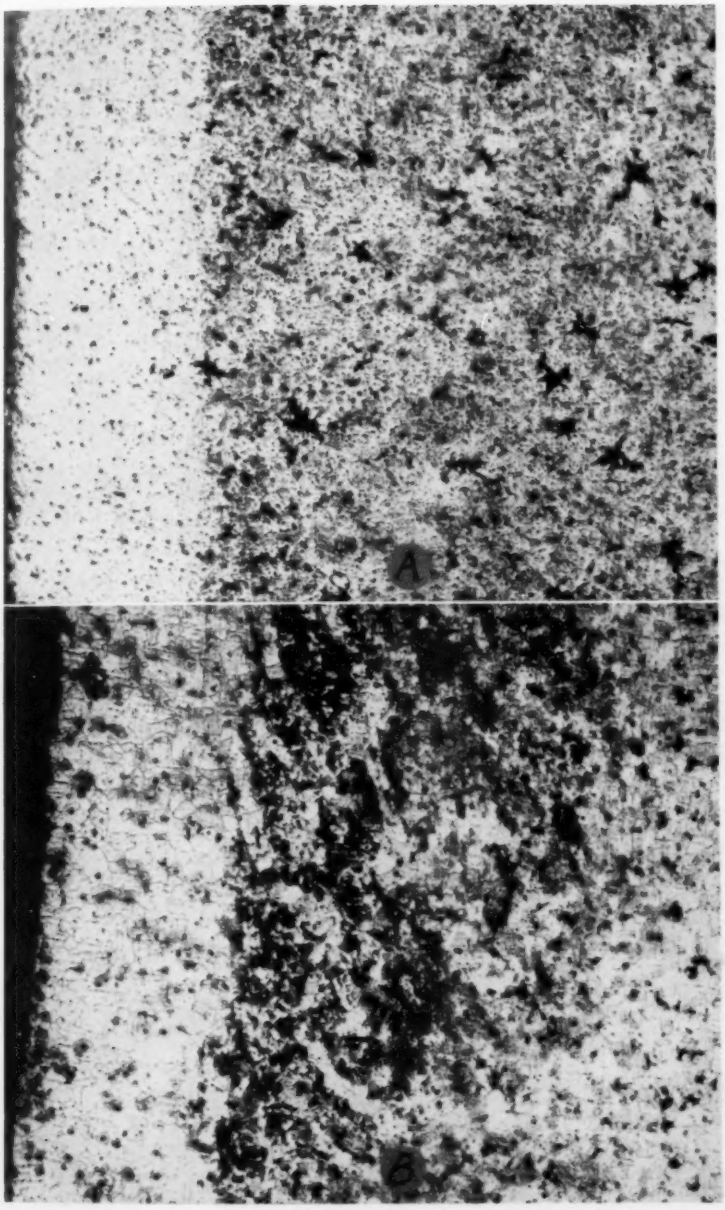

FIG. 3—MICROGRAPHS OF SECTIONS THROUGH AS-CAST (A) AND MACHINED (B) FACES OF SAMPLE NO. 1 HELD FOR A PROLONGED FIRST STAGE AND ACCELERATED SECOND STAGE ANNEAL IN A DECARBURIZING ATMOSPHERE. PEARLITE EXTENDS DEEPER UNDER AS-CAST FACE.—80X.

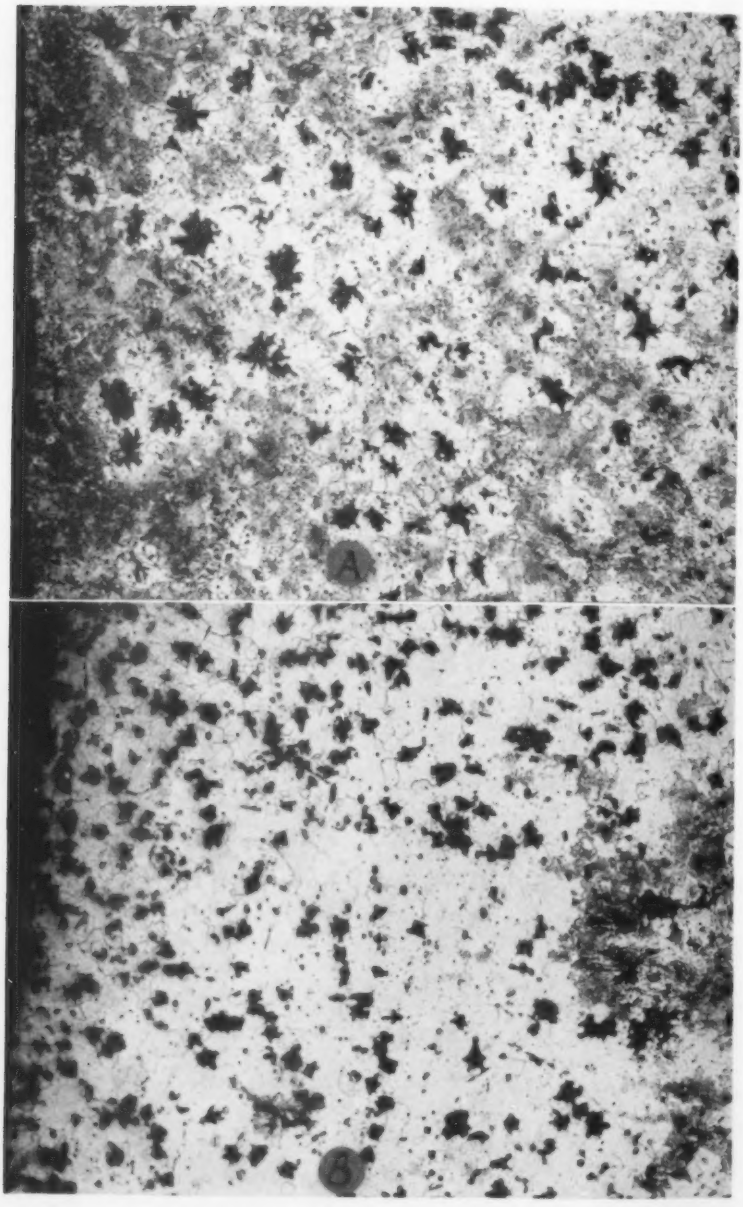

FIG. 4—MICROGRAPHS OF SECTIONS THROUGH AS-CAST (A) AND MACHINED (B) FACES OF SAMPLE NO. 3 HELD FOR PROLONGED FIRST-STAGE AND ACCELERATED SECOND STAGE ANNEAL IN A CARBURIZING ATMOSPHERE. NOTE ALMOST COMPLETE ANNEAL UNDER MACHINED FACE AND HEAVY PEARLITE UNDER AS-CAST FACE.—80X.

20. The bars annealed in a carburizing gas were all held for their stage annealing in a gas carburizing furnace in which the atmosphere consisted of approximately 85 per cent charcoal gas and 15 per cent natural gas. After the high temperature treatment, only the bar from which samples 3 and 4 were made, showed no ferrite around the carbide particles for a depth of 0.010 in. from the machined surface and 0.050 in. from the unmachined faces.

21. Samples 3 and 4 were almost identical in so far as relative annealing rates are concerned. Neither showed any positive evidence of carburizing action during the second stage as would be indicated by pearlite to a depth of 0.005 in. or less on the machined face. Both samples were as far progressed in annealing at the machined faces as in the core of the casting. If any part seemed more nearly completely annealed it was near the surface. However, under the as-cast faces, heavy pearlite is visible to a depth of 0.100 in., the last 0.030 in. of which has gradually less pearlite finally becoming similar to the average core structure. Photo micrographs of areas next to the machined and as-cast surfaces representative of these samples are reproduced in Fig. 4.

22. This phenomenon, of sub-surfaces of machined and cast faces showing very different annealing characteristics, is apparent, irrespective of the type of gas used. It is either inherent in the casting or results from an action of both types of gas used in the tests because of a difference in the characteristics of the surfaces or of the sub-surface metal.

23. As a check on this, a bar with one face machined was annealed with production castings in a furnace using a decarburizing atmosphere. Micro-samples from this bar showed only 0.002 in. total decarburization, backed by 0.010 in. of pearlite on the machined face and 0.010 in. of total decarburization backed by approximately 0.013 in. of pearlite on the as-cast face. That is, pearlite appeared to a depth of 0.012 in. under the machined face and to a depth of 0.023 in. under the as-cast face. Photo micrographs of this sample are reproduced in Fig. 5.

24. The total decarburization being less on this bar than on samples 1 and 2 is probably due to the shorter time in the first stage of annealing. The bar was held approximately 10 hr. at 1750°F. as against 24 hr. on samples 1 and 2. The shallower depth of pearlite may also be due to this fact, but probably is more

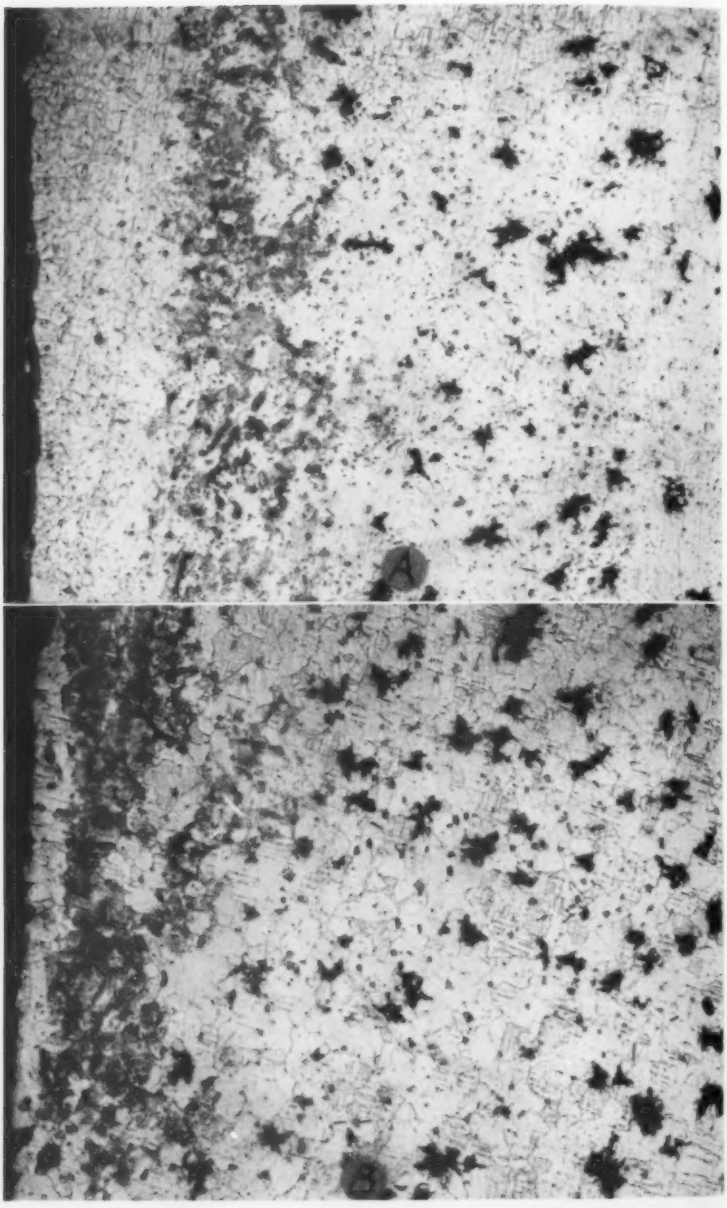

FIG. 5—MICROGRAPHS OF SECTIONS THROUGH AS-CAST (A) AND MACHINED (B) FACES OF SAMPLE ANNEALED WITH PRODUCTION IN A FURNACE HAVING A DECARBURIZING ATMOSPHERE. NOTE DECARBURIZATION AND RETAINED PEARLITE EXTEND DEEPER UNDER AS-CAST FACE THAN UNDER MACHINED FACE.—80X.

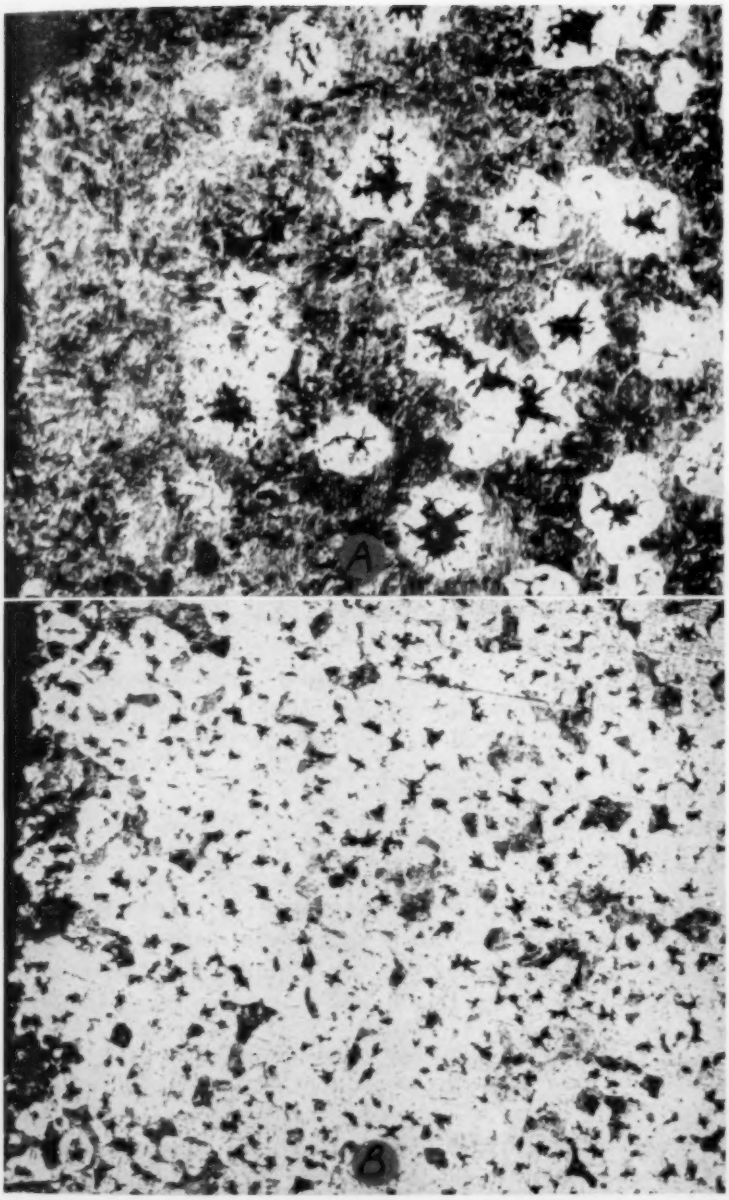

FIG. 6—MICROGRAPHS OF SECTIONS THROUGH AS-CAST (A) AND MACHINED (B) FACES OF SAMPLE HELD 72 HOURS AT 1680°F. IN A CARBURIZING ATMOSPHERE FOLLOWED BY 18 HOURS AT 1450°F. WITH ALL FACES PROTECTED. NOTE IRON UNDER MACHINED FACE AND IN CORE ALMOST COMPLETELY ANNEALED AND HEAVY PEARLITE UNDER AS-CAST FACE.—80X.

likely due to the second stage of annealing being 10 hr. rather than 5 hr.

25. Another sample of high silicon iron treated in the carburizing atmosphere behaved similarly to samples 3 and 4. A third sample of a different heat of iron similar in analysis to samples 3 and 4 was held 72 hr. in a carburizing atmosphere at 1680°F. After cooling it was reheated to 1450°F. and held with its surfaces protected for 18 hr. This sample showed that both the machined face and core were almost completely annealed whereas excessive pearlite appeared to a depth of 0.120 in. under the as-cast face. Photo micrographs of this sample are reproduced in Fig. 6.

26. These tests seem to indicate that a carburizing gas atmosphere, if anything, slightly accelerates rather than retards the annealing of malleable iron whereas a decarburizing atmosphere does have a marked undesirable effect. However, before this can be considered a fact, a further series of tests are indicated in which part of each test bar treated is protected from the action of the gas by an impervious coating of some kind in the *first* stage annealing as well as during the accelerated second stage of the cycle. A comparison of these samples will show definitely whether the action of a carburizing gas improves or retards the annealing rate of the as-cast sub-surface and whether the phenomenon shown above is due to an inherent characteristic of the white iron casting or partially at least due to gas action.

APPLICATION OF ATMOSPHERE GASES TO FURNACES

27. The as-cast sub-surface apparently anneals more slowly than the core metal in both types of atmosphere tested. For this reason the time of the annealing cycle must be based on the annealing rate of this sub-surface metal.

28. Where the time cycle is adequate for this slower annealing part of the casting, the avoiding of decarburization is the main consideration in the application of atmosphere gas to a furnace. If there is any beneficial action at all from a carburizing gas on an as-cast sub-surface metal, it is apparently very slight in terms of the difference in the annealing rate of this metal and the core. For this reason, a carburizing gas is not indicated, provided decarburization can be avoided when bright annealing gas is used.

29. The annealing box points the way to the best method of applying bright annealing gas to a furnace and at the same time avoiding decarburization. In a well luted annealing box, there is no active movement or flow of gas over the work being annealed. Nor is the volume of gas in the box, or that allowed to enter the box, great enough to have any material effect on the load of iron. Without any material change in the iron, the small volume of gas in the box is rapidly changed to one which is neutral in its action to all elements in the iron.

30. As in the case of a box, a furnace must be almost perfectly gas tight if this principle is to be applied. Infiltration of air through small cracks in the furnace structure will cause decarburization and the accompanying untransformed pearlite behind it.

31. Applying the principles of the box to a furnace consists of using only sufficient gas to prevent air infiltration and feeding this gas to the furnace as near as possible to where it escapes. In the case of a continuous furnace almost all the gas should be fed to the vestibules which necessarily must be flushed between pushes. In the case of batch furnaces, the doors can be made gas tight and almost no gas is required.

32. The term "almost perfectly gas-tight furnace" is a relative one, and can best be defined by a consideration of the amount of gas which may escape through small cracks. Any furnace when heated will act like a low, but very efficient chimney. There is a strong tendency for gas to leak out the top of the furnace, and for air to seep in at the bottom. There is normally enough pressure at the top of a furnace to force all the gas, which should be fed to a large continuous furnace, out through a 1 in. diameter hole. It is very easy for a number of small cracks to add up an area greater than this figure. Air infiltration inevitably follows.

33. Possible furnace tightness should be considered of major importance when selecting the type of batch furnace best suited for malleable iron annealing without pots. Muffles, unless gas-tight, will introduce operating problems. Doors should be provided with gaskets. Car type furnaces are difficult to seal. Tube fired or electric furnaces with trays operating on rollers and the bell or cover type of furnace can easily be made gas tight on the other hand. Where unavoidable leakage occurs, the addition of a

hydrocarbon to the gas fed to the furnace will reduce decarburization and the effects of it on the annealing of the iron.

34. Some production annealing furnaces, on the other hand, are being purposely operated to maintain a highly active decarburizing gas within the furnace chamber. A very lean gas-air mixture is fed to the generator, in order to make a gas that will remove a maximum of carbon from the iron per unit volume.

35. This type of operation has in certain instances given better iron than when a more neutral atmosphere was maintained in the furnace. The reason for this is apparently that in a neutral atmosphere there was too much dark-etching pearlite remaining in the slower annealing metal immediately under the surface of the castings. This pearlitic layer would probably have annealed with a somewhat longer time cycle. However, instead of increasing the time, the combined carbon in the pearlite was reduced by decarburization rather than graphitization. Satisfactory commercial results are obtained if this decarburization proceeds to a point where the pearlite is in patches and etches rather light, even though it is noticeable to a depth of 0.030 in. When this principle of operation is being applied, large volumes of lean-mixture bright-annealing gas should be fed to the furnace at points where the gas will be made to flow over the iron in the high temperature part of the furnace.

36. Pearlitic malleable iron presents a somewhat different problem. Many parts are to be hardened after grinding operations from which very little stock is removed. For this reason decarburization during the annealing cycle can not be tolerated.

37. The best operating combination for pearlitic malleable should be charcoal gas applied to a tight furnace. In this instance the gas may be fed to the furnace at any point.

38. Another method, applicable only to a tight furnace, is to flush the vestibules with bright-annealing gas and feed into the hot zone of the furnace a very small volume — e. g. 100 to 200 cu. ft. per hr. — of bright-annealing gas to which has been added approximately 15 per cent of its volume of natural gas. Large volumes of natural gas and bright-annealing gas mixtures fed to a furnace will cause operating troubles due to the accumulation of carbon in the furnace and on the heating tubes.

ACKNOWLEDGMENT

39. The writer would like to express his thanks for the help given him by Messrs. Joseph and Clark, Saginaw Malleable Iron Division, General Motors Corporation, Mr. Boegehold, General Motors Research Laboratories, and Messrs. Ruckstahl and Miller, Holcroft and Company.

DISCUSSION

Presiding: C. L. JOSEPH, Saginaw Malleable Iron Division, General Motors Corporation, Saginaw, Michigan.

F. L. WOLF[1]: What are the commercial possibilities of using this process in an intermittent type annealing furnace? I assume that the data presented were obtained on an electric furnace. What are the commercial possibilities of applying gas control to the intermittent type furnace? Considering a coal-fired furnace what could be done to make use of atmosphere control?

MR. DOW: We have had no experience using powdered coal in atmosphere furnaces. Gas is, of course, being quite commonly fired in radiant tubes that act in a furnace like heating elements. Tests have been made with powdered coal fired in radiant tubes.

MR. WOLF: I was thinking about the ordinary intermittent type of annealing furnace in which you put your charge, bring it up to your holding temperature and hold it for 40, 50 or 60 hours, whatever the cycle may be. If the furnace is powdered coal fired, oil fired or gas fired it is necessary to remove the products of combustion frequently so it appears as if it would be impractical to control the furnace atmosphere. If practical it would probably be an expensive procedure.

CHAIRMAN JOSEPH: I would think the way to do that, would be to install radiant tubes. I don't believe you can get atmosphere control in a powdered coal fired unit.

MR. DOW: I believe the first requisite of the gas would be almost impossible to obtain; i. e., that it be non-scaling. The coal would have to be fired on the outside of a muffle, and I think it would be very difficult to keep a refractory muffle free of those small cracks which can cause so much damage.

C. C. LAWSON:[2] We have a radiant tube furnace and we have a gas machine practically like Mr. Dow showed on his first chart. We run this oven at 1100 to 1200°F. and purge the furnace, and we hold the CO at about 15 per cent and the CO_2 usually runs 1 to 2 and some-

[1] Technical Supt., Ohio Brass Co., Mansfield, Ohio.
[2] Superintendent, Wagner Malleable Iron Co., Decatur, Ill.

times 3 per cent, and we must have all cracks in the oven sealed. We can't have any leaks in the tubes because if we do, we get into trouble. In other words, we lose our pressure right away.

We find that most of our pearlitic iron is caused from cooling rather than from the mixture of gases that we have in this furnace during the cycle. In other words, we hold it right under the critical point for about 8 hours and we find that this eliminates all traces of pearlite, either on the edges or elsewhere through the casting. We think that the cooling process of this oven is more important, really, than the gases that are fed into it. Of course, we know that the gases have to be neutral, or practically so. We do find that the cooling is where we do or do not get pearlite.

I don't believe I can add much on the gases to what Mr. Dow has said. I think he covered that pretty thoroughly. We have operated this furnace for about 2½ years and just lately we have found that we had to go in and take some of the tubes out. The tubes were screwed together in sections and we found that some of the tubes had started to pull and loosen up around the threads. We have taken these tubes out and welded around the threads and put them back in, and I think the company that built this oven is now welding all their tubes that go in instead of screwing them in in sections.

CHAIRMAN JOSEPH: What you are doing, you are holding for 8 hours rather than to control the gas close enough to prevent the rim.

MR. LAWSON: We are holding the atmosphere in there to control the rim all right. I don't mean that we have an oxidizing condition in there at all. But I do know that, regardless of how you hold your atmosphere in there, if you don't do it right, you will get pearlite next to the surface.

MR. DOW: That seems to be the experience of others — that the action of the gas is a factor of minor importance in the formation of the pearlitic rim, unless the gas be actively decarburizing and the flow be high. The analysis of the gas given by Mr. Lawson is that of an almost neutral gas, very, very slightly decarburizing if any.

MR. WOLF: What about the extra expense using gas control? How much does it add to the expense of annealing?

MR. LAWSON: We cut our annealing cost one-third from our old procedure in gas and we get this natural gas pretty cheaply down in Decatur. We pay 21 cents for it and that is a lot cheaper than most can buy gas, I think.

CHAIRMAN JOSEPH: We use around 2800 cubic feet of natural gas per ton and about 400 cubic feet of that goes to the control of the atmosphere.

H. BORNSTEIN[3]: Our work has been done on a one-ton radiant tube furnace, more or less experimental, and our experience is just about the same as that of the Wagner Malleable. We have found the annealing cycle is much more important than the analysis of the gas, when we get the gas anywhere near right. We are using a mixture of natural gas and partially burned (DX) gas and we obtain a furnace atmosphere similar to that described by Mr. Lawson. For further details see our discussion of paper by De Coriolus and Cowen in the 1938 Transaction of A. F. A., pages 13-17. When our cooling rate is too fast during the secondary stage of the annealing, we will get pearlite. We can eliminate it by slowing down.

MR. DOW: It is my belief that this tendency for many irons to anneal more slowly under the cast surface than in the core has been wrongly blamed on the gas atmosphere, much of it that is not due to the gas atmosphere at all.

MR. BORNSTEIN: I might say that we tried out a number of combinations of atmosphere and we found we could vary the atmosphere over a pretty wide range without, seemingly, any effect on the amount of decarburization or the amount of pearlite that we got.

CHAIRMAN JOSEPH: A short time ago we took some white iron bars, ground one-quarter inch off the surface and packed some of these bars in a pot and sealed the pot tight. We took the same iron and ground the bars about one-quarter inch deep and put them through the atmosphere control oven and found that the bars which were sealed in the can and put through the oven, on both the machines and the as-cast surface, showed no difference. The temper carbon spots were right out to the outside of the casting. The bars put into the oven without being sealed showed the temper carbon spots further in from the surface, with a far less depth of decarburization on the machined surface than on the as-cast surface, which would indicate that the atmosphere is important, because in the sealed can, the atmosphere didn't have any hydrocarbons present, whereas in the open atmosphere in the oven, the hydrocarbons possibly attacked the surface of the metal and were the cause of the heavy decarburization on the as-cast surface, whereas on the machined surface the decarburization was very low and the temper carbon spots were practically out to the surface. So I still think that the atmosphere is very important. But I do think that there are other things besides carbon monoxide and carbon dioxide that we can't really watch very closely that causes this.

E. G. DE CORIOLUS[4]: I am inclined to agree with those who say that the atmosphere in the malleablizing furnace is not of such tremendous importance. In other words, to be sure, you want an atmosphere which is non-scaling to begin with, but so far as rigidly balancing your atmosphere with respect to the metal, I believe you are attempting to do something which is not absolutely essential for a commercial product.

[3] Director of Laboratories, Deere & Co., Moline, Ill.
[4] Research Director, Holcroft & Corp., Toledo, Ohio.

We are operating a continuous malleablizing furnace today with the atmosphere secured directly from the burner equipment of the furnace itself. We have no external preparation unit of any kind and we make up for deficiencies of carburizing agent by adding a very small amount of charcoal to each load as it goes through the furnace. We have a slightly decarburized rim which is no greater than one which we secure from other furnaces where we attempt to more closely control the atmosphere.

I have had no experience in determining the rate of decarburization of the as-cast bar as compared to the machined bar. The work which Mr. Dow has done on this is very interesting. But from the commercial standpoint, such as is now required, at any rate, it seems to me that we don't want to build up in the minds of the foundrymen the idea that the atmosphere is of tremendous importance, but that it is something which is very easily controllable and that you don't have to fuss too much about the operation of your furnace. The flue gases that you can produce by the combustion of the usual natural gas are perfectly satisfactory and, as a matter of fact, for the addition of your non-decarburizing constituent, a small amount of charcoal will go a long way to help out in the operation of the furnace.

MR. LAWSON: I don't want anybody to get the impression I made the statement that we do not control our atmosphere. I wasn't trying to detract at all from the neutral atmosphere, because we do control our atmosphere.

CHAIRMAN JOSEPH: We understand you do, except we were glad to hear that you don't think the atmosphere is as important as a lot of people think it is; but I still think it is very important, because in the summer time when the moisture content of the atmosphere is high, we find that the decarburization is greater than it is in the winter time, and there are very few of us who take care of the moisture in the atmosphere that goes into the radiant tube ovens. All we do is to cool the gas down to the temperature of the room, which is, in the summer time, rather high, and the atmosphere contains a lot of moisture, whereas in the winter time the atmosphere is colder and there is less moisture in the air. But there is something there that causes more decarburization due to the moisture content. It is probable that the moisture breaks up into hydrogen and oxygen, and causes a greater amount of decarburization when the moisture content is high.

H. A. SCHWARTZ[5]: I recall some 15 years ago when we were playing with these things on a laboratory basis that we were interested in the reaction of hydrogen gas with white cast iron, the experiment being merely a combustion furnace with a quartz tube in it, some pulverized white cast iron in the bottom of the inside and a stream of hydrogen passing through. The experiment was not originally set up to see how much carbon would come out. We very promptly found that if we burned the gases coming out of this tube, burned copper oxide and passed the

[5] Manager of Research, National Malleable & Steel Castings Co., Cleveland, Ohio.

products of combustion through a barium hydroxide solution, we got a lot of barium carbonate. Our first thought was that our hydrogen, which was tank hydrogen, was not very pure, and we were getting a reaction with traces of oxygen.

We finally set that experiment up, purifying the hydrogen of oxygen by passing it through the liquid sodium potassium alloy, which we felt sure was a better reducing agent than iron could be, and we still had a lot of presumably methane, maybe one of the other methane hydrocarbons coming out. That reaction will continue down to temperatures of just a very few hundred degrees centigrade. My recollection from memory is that we found it as low as 200°C., certainly as low as 300 to 400°C.

So there are other means of decarburization as well as oxidation, and the reaction which combines the hydrogen, and oxygen to produce methane, or some of the methane-propane series, is quite important. How important it is to the malleable man who wants to anneal some iron may be another question. I offer that comment on hydrogen merely in support of the Chairman's comments rather than as having to do with what you get in an actual annealing furnace.

W. D. MCMILLAN[6]: We have had about three years' experience with the equipment and there are two items that I think I ought to mention. We have mentioned the fact that the atmosphere is not important, that the cooling rate is more important, but I think that we should qualify that a little bit by saying that the atmosphere is very important from a non-scaling standpoint but from the control of the pearlitic skin it is less important. However, we have had this experience, that the volume of gas that goes through the furnace will have a direct effect on the amount of pearlitic skin we get. We run about 300 or 400 cubic feet of gas per hour, that is, atmospheric gas, and if we keep that down, we have a much thinner pearlitic skin than if there is a larger volume of gas that runs through the furnace. Now, that indicates there is to some extent a quantitative reaction. The more gas, the more skin. Therefore, there must be some relationship between the pearlitic skin and the amount of gas that is used.

MR. DOW: In our tests even the machined surfaces showed a retarded annealing rate where the action of the atmosphere was decarburizing and it is my belief that an actively decarburizing gas will cause pearlite to appear under the surface of the iron where the annealing time has been adequate to fully anneal the iron, were the decarburization absent. With an almost neutral atmosphere this becomes a minor affect and very little extra second stage annealing time is sufficient to eliminate the rim.

It is easy to get a neutral atmosphere in a tight furnace but where leaks require that a large volume of gas be fed to the furnace to prevent scaling it is very difficult to avoid decarburization and the deeper pearlitic edge that results.

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[6] Metallurgist, International Harvester Co., Chicago, Ill.

This paper was first prepared some time ago with all tests on unmachined surfaces. The results indicated there was some factor other than atmosphere affecting the surface rate of annealing and the tests were all re-run with machined and unmachined surfaces for comparison.

MEMBER: I would like to know what is meant by a machined surface. The author is talking about the machined surface of white iron and what is the difference between a machined and an un-machined surface from the standpoint of the reaction in the annealing oven? I just don't follow that trend of thought.

MR. DOW: In our earlier tests, irrespective of the atmosphere, there seemed to be a slower annealing rate under the as-cast surface, enough so that we felt there was something there other than the atmosphere affecting the annealing. The core seemed always to anneal faster. So we machined the surface off the white iron in order to have core iron at a surface where the gas could affect it. This gave us a more equal comparison—identical irons, one affected by the gas and one not. The procedure is of no value from a practical angle but it does show better the affect of the action of the gas.

Properties of Some Copper-Bearing Cast Steels

C. T. Greenidge* and C. H. Lorig,* Columbus, Ohio

Abstract

The effects of up to 2.40 per cent of copper on the mechanical properties of heat treated plain-copper and low-alloy copper cast steels are discussed. The low-alloy copper cast steels contain one or more of the following elements: vanadium, titanium, aluminum, molybdenum, columbium, zirconium, chromium, silicon, and manganese, the last two in higher than normal percentages. Properties of the plain copper cast steels in the fully annealed, the normalized, the normalized and drawn, and the quenched and drawn conditions, and of the low-alloy copper cast steels in the fully annealed, the normalized, and the normalized and drawn conditions are given.

Copper contents above 0.50 per cent increase markedly the tensile and yield strengths of fully annealed, of normalized, and of quenched and drawn cast steels with only a slight lowering of the ductility and resistance to impact. Since the copper increases the yield strength more rapidly than the tensile strength, the copper-bearing steels have high yield ratios. The unusual ability of copper steels containing 0.7 to 2.0 per cent of copper to precipitation harden makes it possible to increase the hardness, tensile and yield strengths without lowering the ductility proportionately by drawing normalized steel at 800 to 1100°F.

Several alloying combinations containing copper and one or more of the other common alloying elements are satisfactory for use in low-alloy, high strength cast steels. Two types of low alloy cast steels are discussed in detail; namely, the copper-manganese-titanium steels and the copper-chromium steels containing either aluminum or titanium.

* Metallurgist, Battelle Memorial Institute.

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INTRODUCTION

1. Copper has been used commercially as well as experimentally in cast steels for the past 15 yrs. Kinnear[1], in 1925, discovered that copper conferred precipitation hardening to steel and utilized this specific action of copper to raise the tensile strength, yield strength, and hardness of commercial steel castings. In copper-steels, the copper enters the ferrite from which all except about 0.35 per cent may separate upon very slow cooling or upon arrested cooling or upon reheating for an extended period of time above 800°F. On air-cooling and even moderately slow cooling, the copper is retained in the ferrite where it functions as an effective ferrite strengthener. In this capacity it is of value in producing cast steels having high yield strength and high yield ratio. The copper content of commercial cast steels rarely exceeds 2.00 per cent.

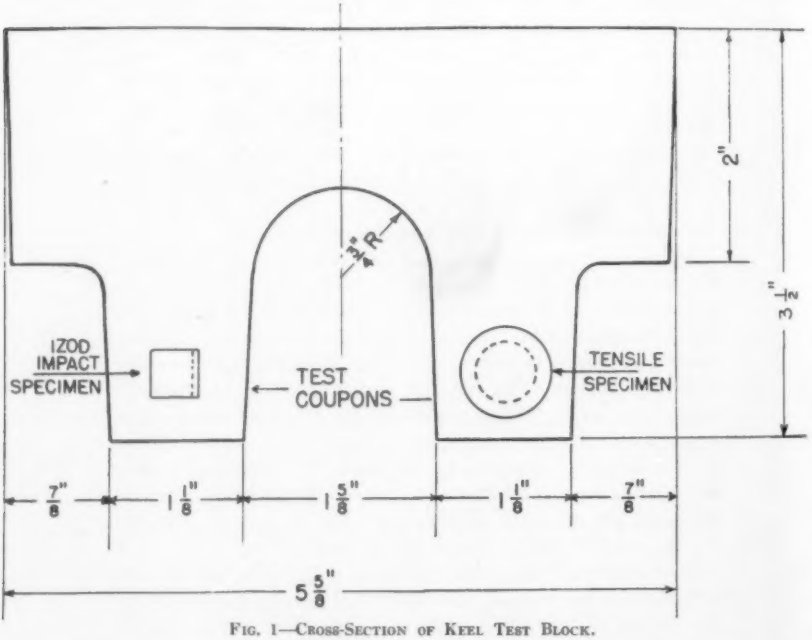


FIG. 1—CROSS-SECTION OF KEEL TEST BLOCK.

2. With the exception of the most recent publications, the literature on copper cast steels is dealt with in "The Alloys of Iron and Copper"[2] and in "Copper in Cast Steel and Iron"[3] and again rather briefly in "Steel and its Heat Treatment."[4] In spite

[1] Superior numbers refer to bibliography at end of paper.

of the treatment in these sources, there is a dearth of information on the effect of copper in cast steels. An investigation designed to provide more information of this kind constituted one phase of a research program sponsored at Battelle Memorial Institute by Anaconda Copper Mining Company, Phelps-Dodge Corporation, and Kennecott Copper Corporation for the benefit of the copper industry. This paper presents some of the results of the investigation.

PLAIN COPPER CAST STEELS

3. The plain-copper steels as well as the low-alloy copper steels were prepared in either an acid or a basic-lined high frequency induction furnace (the steels from either furnace had essentially the same properties) and were poured into 65 lb. double-keel block test castings of a section size shown in Fig. 1. The legs of the block were 15 in. long. Each leg provided material for two tensile specimens and one triple-notched, Izod specimen. The usual procedure was to remove the legs with a torch, heat-treat them and finally machine the specimens.

Fully Annealed Copper Cast Steels

4. Properties of copper steels fully annealed for 1 hr. at 1600 to 1650°F. are given in Table 1 and shown graphically in Fig. 2.

5. Additions of less than 0.50 per cent copper had only a minor effect on the mechanical properties. As the copper content increased from 0.50 to 1.30 per cent, the tensile strength, yield strength, and hardness increased rather rapidly; and the elongation, reduction of area, and impact resistance decreased slowly. Little further change in properties was effected by increasing the copper content from 1.30 to 2.40 per cent. In fact it appears that the tensile strength and hardness reached a maximum value at 2.00 per cent copper.

6. A few tests indicated that the fully annealed copper steels did not respond to a precipitation hardening treatment. This was undoubtedly due to the fact that cooling from the annealing temperature was too slow to retain a sufficient amount of copper to form a supersaturated solution with ferrite. This failure to precipitation harden is characteristic of copper steels on full annealing.

Normalized and Normalized and Drawn Copper Cast Steels

7. The properties of normalized and normalized and drawn copper cast steels are set forth in Table 2. Properties of the steels

Table 1

MECHANICAL PROPERTIES OF FULLY ANNEALED PLAIN COPPER CAST STEELS

Chemical Composition, Per Cent						Heat Treatment, °F.	Mechanical Properties					
C	Mn	Si	P	S	Cu		Tensile Strength, lb. per sq. in.	Yield Strength, lb. per sq. in.	Elongation in 2 in., Per Cent	Reduction of Area, Per Cent	Impact[2] ft.-lb.	Brinell Hardness
LOW CARBON STEELS												
0.19	0.63	0.33	0.012	0.030	.....	1600	62,000	42,000	36.5	59.8	61.0 I	.....
0.20	0.76	0.33	0.017	0.032	.....	N+A	70,000	43,500	33.0	53.8	.....	128
0.16	0.68	0.32	.....	.....	0.89	1600	67,000	49,500	32.5	53.9	23.0 C	119
0.16	0.90	0.40	.....	.....	1.15	N+A	73,250	55,250	32.0	59.0	.....	137
0.15	0.60	0.35	.....	.....	1.66	N+A	76,750	60,500	29.0	48.0	.....	152
0.19	0.63	0.40	.....	.....	1.88	N+A	76,250	58,500	31.0	53.0	.....	152
MEDIUM CARBON STEELS												
0.27	0.79	0.43	.....	.....	.....	1600	74,500	45,250	30.5	48.9	21.0 C	126
0.32	0.80	0.40	0.014	0.031	.....	1650	71,750	42,000	30.0	45.5	26.0 I	.....
0.31	0.78	0.36	.....	.....	0.52	1650	81,750	47,750	27.0	42.2	16.0 C	148
0.30	0.87	0.47	.....	.....	0.78	1600	83,750	56,500	26.5	44.9	18.5 C	142
0.27	1.03	0.47	.....	.....	1.13	N+A	87,750	58,500	27.0	49.0	.....	170
0.29	0.65	0.34	0.014	0.032	1.24	1650	85,000	57,000	25.8	44.3	22.5 I	.....
0.31	0.75	0.42	0.029	0.034	.....	1650	77,000	48,500	28.7	44.5	14.0 C	134
0.31	0.75	0.42	0.029	0.034	1.21	1650	90,000	61,000	24.8	43.6	12.0 C	161

(Continued on next page)

Table 1 *(Continued)*

0.31	0.75	0.42	0.029	0.034	1.76	1650	90,500	58,500	24.5	41.4	10.0 C	159
0.31	0.82	0.44	.....	.....	2.13	1600	86,750	65,750	25.9	42.8	18.5 C	152
0.31	0.75	0.42	0.029	0.034	2.40	1650	89,500	58,500	24.5	44.9	12.0 C	157
0.35	0.75	0.40	.....	.....	0.63	1600	84,000	52,750	26.0	41.2	.....	159
0.37	0.75	0.53	.....	.....	0.82	1600	90,250	60,000	23.5	36.6	14.0 C	156
0.35	0.78	0.44	.....	.....	1.70	1600	90,000	65,750	24.0	39.8	15.0 C	156
0.39	0.95	0.52	.....	.....	.....	N+A	86,250	49,250	26.0	39.0	.....	159
0.40	0.91	0.42	.....	.....	1.24	N+A	96,250	61,750	22.0	35.0	.....	183
						PRECIPITATION HARDENED STEELS						
0.35	0.78	0.44	.....	.....	1.70	1600	90,000	65,750	24.0	39.8	14.5 C	156
0.35	0.78	0.44	.....	.....	1.70	1600; 930[3]	89,800	65,500	22.5	33.8	14.5 C	158
0.31	0.82	0.44	.....	.....	2.13	1600	86,750	65,750	25.9	42.8	18.5 C	152
0.31	0.82	0.44	.....	.....	2.13	1600; 930[3]	87,800	65,750	25.0	42.8	17.5 C	155
						HIGH SILICON STEELS						
0.17	0.67	1.05	0.012	0.029	1.25	1650	80,000	60,500	30.8	56.5	36.5 I	.....
0.17	1.19	1.04	0.012	0.032	1.26	1600	83,500	61,500	29.8	57.0	38.5 I	.....

[1] N+A = 5 hr. at 1650°F., air cooled, 1575°F. for 1 hr. furnace cooled.
All other steels were heated for 1 hr. at the indicated temperatures.
[2] I = Izod impact; C = Charpy impact.
[3] Precipitation hardening treatment 3 hr. at 930°F.

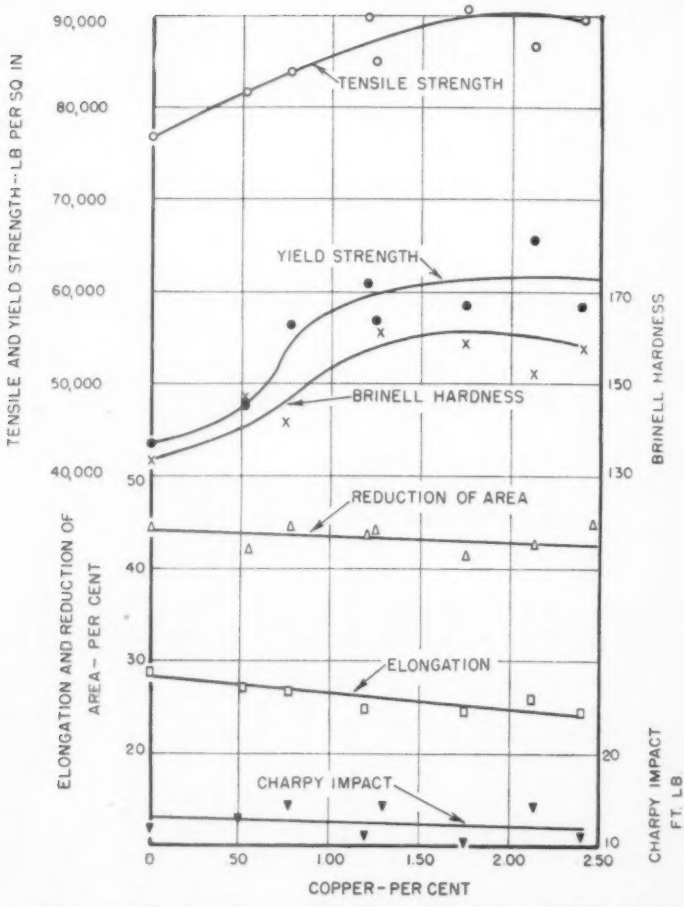


FIG. 2—EFFECT OF COPPER ON PROPERTIES OF FULLY ANNEALED CAST STEELS CONTAINING 0.29—0.31 PER CENT C, 0.65—0.87 PER CENT MN, 0.34—0.44 PER CENT SI.

in this table show that copper contents below 0.50 per cent have little effect on the mechanical properties. Above 0.5 per cent and up to 2.0 per cent copper, the tensile strength, yield strength, and Brinell hardness increase rapidly while the elongation, the reduction of area, and the resistance to impact decrease slowly with increase in copper content. Above 2.0 per cent copper the increase in strength with copper content is small. The effects of copper on properties of normalized cast steels are shown graphically by the curves of Fig. 3.

8. In the normalized steels the copper forms a supersaturated

solution with the ferrite whereas in steels that are fully annealed, copper in excess of the solid solubility of copper in ferrite has apparently coalesced into particles larger than the critical size for precipitation hardening. The copper held in supersaturated solution in the normalized steels strengthens the ferrite more than if it were precipitated in particles larger than the critical size for precipitation hardening; hence its influence in the normalized steels is more pronounced than in the fully annealed steels.

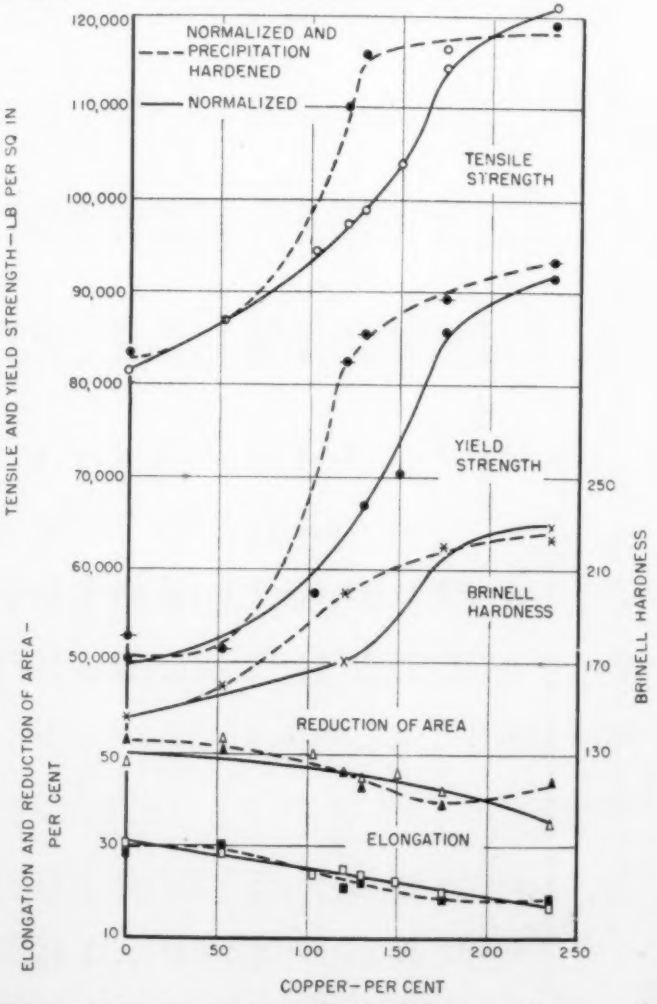


FIG. 3—EFFECT OF COPPER CONTENT AND PRECIPITATION HARDENING ON MECHANICAL PROPERTIES OF NORMALIZED CAST STEELS CONTAINING 0.30—0.32 PER CENT C, 0.65—0.85 PER CENT MN, 0.35—0.45 PER CENT SI.

Table 2

PROPERTIES OF NORMALIZED AND NORMALIZED AND DRAWN PLAIN-COPPER, CAST STEELS

Chemical Composition, Per Cent						Heat Treatment		Mechanical Properties					
C	Mn	Si	P	S	Cu	Normalized. °F.	Drawn °F.	Tensile Strength lb. per sq. in.	Yield Strength lb. per sq. in.	Elongation in 2 in. per cent	Reduction of Area per cent	Impact* ft. lb.	Brinell Hardness
LOW CARBON STEELS													
0.09	0.48	0.23	.....	.....	0.72	1650	.....	58,500	42,000	32.7	50.7	37.5 C	105
0.09	0.48	0.23	.....	.....	0.72	1650	930[2]	61,000	48,000	32.5	52.2	37.0 C	113
0.10	0.50	0.25	.....	.....	1.07	1650	.....	61,500	47,000	32.0	53.7	36.0 C	113
0.10	0.50	0.25	.....	.....	1.07	1650	930[2]	79,500	67,000	23.2	38.5	24.5 C	154
0.08	0.42	0.25	.....	.....	1.38	1650	.....	68,000	54,000	31.0	55.6	35.0 C	125
0.08	0.42	0.25	.....	.....	1.38	1650	930[2]	92,000	78,500	22.5	44.1	18.0 C	183
0.16	0.68	0.32	.....	.....	0.89	1600	.....	72,500	55,500	28.7	50.1	27.0 C	129
0.16	0.68	0.32	.....	.....	0.89	1600	930[2]	89,250	74,000	27.0	54.0	24.0 C	156
0.16	0.92	0.45	.....	.....	1.04	1650[1]	1575	81,000	59,000	32.0	62.0	.....	157
0.16	0.92	0.45	.....	.....	1.04	1650[1]	1575 930[2]	100,500	80,700	24.0	45.0	.....	201
0.15	0.60	0.35	.....	.....	1.16	1650[1]	1575 1250	69,000	52,750	33.0	61.0	.....	134
0.19	0.63	0.35	.....	.....	1.27	1650[1]	1575	86,200	64,000	28.0	50.0	.....	170
0.19	0.63	0.35	.....	.....	1.27	1650[1]	1575 930[2]	107,750	89,250	21.5	45.0	.....	223
0.19	0.63	0.35	.....	.....	1.27	1650[1]	1575 1250[3]	74,000	58,000	33.0	63.0	.....	149
0.17	0.70	0.34	.....	.....	1.55	1600	.....	78,000	64,000	28.0	50.6	28.5 C	150
0.17	0.70	0.34	.....	.....	1.55	1600	930[2]	95,000	83,000	17.5	31.2	19.0 C	192
0.15	0.60	0.35	.....	.....	1.66	1650[1]	1575	85,250	64,500	29.0	55.0	.....	170
0.19	0.63	0.35	.....	.....	1.88	1650[1]	1575	91,000	69,000	27.0	54.0	.....	187
0.19	0.63	0.35	.....	.....	1.88	1650[1]	1575 1250	77,500	64,000	27.5	46.0	.....	149
0.16	0.51	0.40	.....	.....	1.92	1600	.....	81,000	68,000	24.5	43.4	18.5 C	156
0.16	0.51	0.40	.....	.....	1.92	1600	930[2]	98,000	85,000	20.0	38.3	11.0 C	190

(Continued on next page)

Table 2 *(Continued)*

MEDIUM CARBON STEELS

0.24	0.65	0.35	0.011	0.032	1.26	1650	.....	.....	83,000	59,500	24.0	56.0	34.0 I	.....
0.24	0.65	0.35	0.011	0.032	1.26	1650	.....	1050	95,000	73,500	22.5	54.9	20.5 I	.....
0.24	0.65	0.35	0.011	0.032	1.26	1650	.....	950[2]	105,000	79,500	22.0	49.5	8.0 I	.....
0.28	0.64	0.34	.....	.....	.....	1700	1600	1200	69,000	43,500	37.8	63.3	45.5 I	.....
0.31	0.75	0.42	0.034	0.024	.....	1650	.....	.....	85,500	54,500	29.5	53.4	21.0 C	146
0.31	0.75	0.42	0.034	0.024	.....	1650	.....	930[2]	83,500	53,000	29.3	51.9	20.0 C	146
0.32	0.80	0.40	0.014	0.031	.....	1650	.....	.....	81,500	50,500	30.5	48.9	29.0 I	.....
0.31	0.78	0.36	.....	.....	0.52	1600	.....	.....	87,000	51,500	29.0	51.7	31.0 I	160
0.31	0.78	0.36	.....	.....	0.52	1600	.....	930[2]	87,000	51,250	30.0	51.0	22.5 C	159
0.30	0.68	0.41	0.012	0.025	0.90	1700	1600	1200	79,500	53,000	28.0	57.0	33.0 I	.....
0.31	0.83	0.36	0.011	0.025	1.03	1700	1600	.....	94,500	57,500	24.6	50.0	.....	.....
0.31	0.83	0.36	0.011	0.025	1.03	1700	1600	1200	84,500	51,000	28.0	59.0	34.0 I	.....
0.31	0.75	0.42	0.034	0.029	1.21	1650	.....	.....	97,500	68,000	25.5	51.0	17.0 C	170
0.31	0.75	0.42	0.034	0.029	1.21	1650	.....	930[2]	110,000	82,500	21.0	45.8	11.5 C	201
0.32	0.78	0.37	0.009	0.028	1.22	1650	.....	.....	96,500	64,500	25.0	49.7	20.0 I	.....
0.32	0.78	0.37	0.009	0.028	1.22	1650	.....	930[2]	116,500	84,000	20.0	43.4	.....	.....
0.32	0.78	0.37	0.009	0.028	1.22	1650	.....	1150	94,000	67,000	21.9	46.8	23.0 I	.....
0.29	0.66	0.34	0.014	0.032	1.24	1700	1600	1200	80,000	58,000	30.3	54.7	36.0 I	.....
0.30	0.83	0.39	0.014	0.031	1.30	1650	.....	.....	99,000	67,000	23.5	44.7	19.5 I	.....
0.30	0.83	0.39	0.014	0.031	1.30	1650	.....	950[2]	116,000	85,500	21.0	43.1	10.0 I	.....
0.30	0.83	0.39	0.014	0.031	1.30	1650	.....	1150	94,500	65,500	25.0	52.2	24.0 I	.....
0.32	0.82	0.36	0.012	0.024	1.50	1700	1600	.....	104,000	70,500	22.0	46.6	26.0 I	.....
0.32	0.82	0.36	0.012	0.024	1.50	1700	1600	1200	89,000	59,000	25.0	55.0	36.0 I	.....
0.31	0.75	0.42	0.034	0.029	1.76	1650	.....	.....	114,500	86,000	19.8	42.0	11.0 C	216
0.31	0.75	0.42	0.034	0.029	1.76	1650	.....	930[2]	116,500	89,500	18.8	38.8	10.0 C	215
0.31	0.75	0.42	0.034	0.029	2.40	1650	.....	.....	121,000	91,500	16.8	34.7	7.0 C	229
0.31	0.75	0.42	0.034	0.029	2.40	1650	.....	930[2]	119,000	93,500	19.3	44.2	9.0 C	223

[1] Heated at 1650°F. for 5 hr.; then cooled in air.
[2] Reheated at 930 to 950°F. for 3 hr.; then cooled in air.
[3] Reheated for 2 hr. at 1250°F. In all of the other cases the steels were heated for 1 hr. at the temperatures indicated.
[4] I = Izod; C = Charpy.

9. Since the normalized steels with more than 0.35 per cent copper, contain copper in supersaturated solution in ferrite, they respond readily to precipitation hardening. The properties of normalized copper steels after a precipitation hardening or drawing treatment of 3 hrs. at 930°F. are also shown in Fig. 3. From this figure it is evident that within the range of about 0.50 to 1.75 per cent copper, precipitation hardening increased the tensile and yield strengths by as much as 16,000 and 20,000 lb. per sq. in. respectively. Maximum effect was obtained at around 1.25 per cent of copper. A moderate decrease in reduction of area and elongation and about a 50 per cent decrease in impact resistance accompanied the increase in tensile and yield strengths. The precipitation hardening effect diminished when the copper content exceeded 1.25 per cent and practically disappeared in the neighborhood of 2 per cent of copper.

10. The effective precipitation hardening temperature range lies between 800 and 1100°F. At the lower temperature or at 800°F., the time required to hold the steel at the precipitation hardening temperature to produce the maximum hardening effect is about 25 hr. The holding time diminishes to about 2 to 3 hr. at 950°F. and to ½ to 1 hr. at 1050°F. Temperatures on either side of the precipitation hardening range affect the copper steels in much the same manner as they do the copper-free steels. The effect of tempering or drawing temperature over the range from 750 to 1300°F. and the effect of tempering at these temperatures for periods from 1 to 6 hr. on a group of 1 per cent copper cast steels containing somewhat less than 0.3 per cent carbon with high and low manganese contents, 0.8 per cent of chromium, 1.5 per cent of nickel, and 1.5 per cent of nickel together with 0.5 per cent of chromium was shown recently by Alexander.[5] He pointed out that due to the lowering of the impact resistance it is not always possible to take advantage of the full increase in strength which can be obtained by precipitation hardening. However, with a steel not too high in carbon and by use of a precipitation temperature higher than that giving maximum hardness very good impact resistance can be obtained in the precipitation hardened copper cast steels.

11. A unique characteristic of the normalized copper cast steels is that the yield strength remains at a high level after drawing at high temperatures. This characteristic together with the precipitation hardening characteristic make it possible to obtain a variety of properties by simply varying the drawing treatment.

Table 3

MECHANICAL PROPERTIES OF QUENCHED AND DRAWN COPPER CAST STEELS

Chemical Composition, Per Cent						Heat Treatment		Mechanical Properties					
C	Mn	Si	P	S	Cu	Quench °F.	Draw °F.	Tensile Strength lb. per sq. in.	Yield Strength lb. per sq. in.	Elongation in 2 in., Per Cent	Reduction of Area, Per Cent	Impact* ft.-lb.	Brinell Hardness
OIL QUENCHED STEELS													
0.32	0.80	0.40	0.014	0.031	.....	1650	1200	83,500	58,500	29.0	60.4	51.0 I	.....
0.30	0.83	0.39	.....	.....	1.30	1650	1200	99,000	70,500	25.0	55.4	50.0 I	.....
0.32	1.52	0.40	.....	.....	1.28	1650	1200	111,500	94,500	21.0	53.4	35.0 I	.....
WATER QUENCHED STEELS													
0.16	0.68	0.32	.....	.....	0.89	1600	1200	83,750	68,250	23.0	48.9	24.5 C	156
0.16	0.90	0.40	.....	.....	1.15	N+QD		84,700	71,000	24.0	51.0	.....	187
0.27	0.79	0.43	.....	.....	.....	1600	1200	90,750	67,000	27.0	55.5	24.5 C	156
0.31	0.75	0.42	0.034	0.029	.....	1650	1200	92,500	66,500	26.0	61.8	32.5 C	164
0.31	0.78	0.36	.....	.....	0.52	1600	1200	98,000	69,125	26.2	57.0	34.0 C	183
0.31	0.75	0.42	0.034	0.029	1.21	1650	1200	104,500	89,000	20.5	55.6	26.0 C	198
0.31	0.75	0.42	0.034	0.029	1.76	1650	1200	113,000	99,000	18.8	46.9	20.5 C	216
0.31	0.75	0.42	0.034	0.029	2.40	1650	1200	116,500	105,000	20.0	52.3	21.5 C	222
0.35	0.80	0.40	.....	.....	0.63	1600	1200	102,750	76,500	19.5	42.2	20.0 C	200
0.35	0.78	0.44	.....	.....	1.70	1600	1200	114,500	102,500	19.5	42.4	17.0 C	221

NOTES: Unless noted otherwise steels were heated at the indicated temperatures for 1 hr.
N+QD = Normalized 1650°F. for 5 hr. 1 hr. at 1575°F., and then water quenched. Drawn for 2 hr. at 1250°F.
* C = Charpy, I = Izod.

Quenched and Drawn Copper Cast Steels

12. The properties of oil quenched and water quenched copper cast steels are shown in Table 3. That the steels have excellent properties when oil quenched and drawn is apparent from the properties of the two steels shown in the table. These steels combine high strength with high ductility and resistance to impact.

13. Properties of some of the water quenched and drawn steels are plotted in Fig. 4. Again, it should be noticed that there

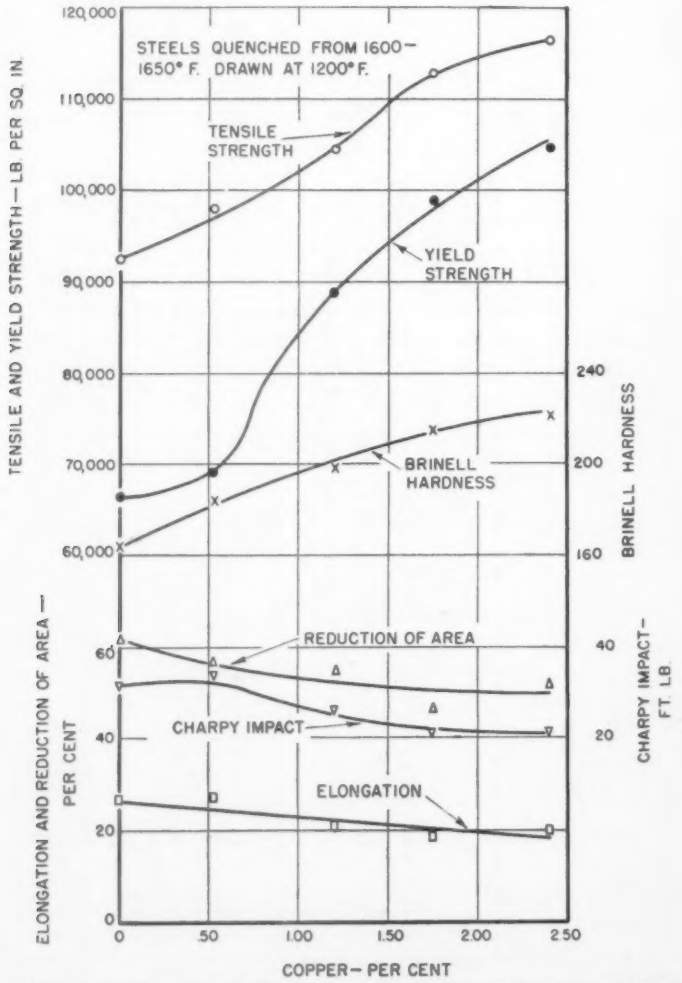


FIG. 4—MECHANICAL PROPERTIES OF WATER QUENCHED AND DRAWN COPPER CAST STEELS CONTAINING 0.30—0.35 PER CENT C, 0.75—0.80 PER CENT MN, 0.40—0.45 PER CENT SI.

is little change in properties up to 0.50 per cent of copper. Above 0.50 per cent of copper and up to at least 2.40 per cent of copper the tensile and yield strengths and the hardness increase, the yield strength increasing much more rapidly than the tensile strength, while the reduction of area, elongation and impact resistance decrease at a relatively moderate rate. The more rapid increase in yield strength as compared with the tensile strength results in steels with increasingly higher yield ratios.

14. Although there are no data given to show the effect of drawing the quenched steels within the precipitation hardening range, past experience with copper steels has shown that any precipitation hardening effects are masked by the softening action of the drawing treatment.

LOW ALLOY COPPER CAST STEEL

Fully Annealed Low Alloy Copper Cast Steel

15. In Table 4 are listed the mechanical properties of a group of fully annealed cast steels containing copper and one of the following elements: vanadium, titanium, aluminum, mobybdenum, columbium, zirconium, and chromium. A few steels in this group have higher than normal percentages of manganese and silicon.

16. Only a few of the elements, manganese, silicon, chromium, and vanadium in the amounts used increased the strength of the fully annealed copper cast steels. The greatest effect of the added elements was on the ductility and impact resistance. Vanadium, titanium, aluminum, zirconium, and chromium appeared to increase the ductility and impact whereas manganese above 1.20 per cent in 1.20 per cent copper steel decreased the ductility and impact. Columbium also decreased both of these properties. As the full benefit of most alloying elements is seldom attained in annealed steels, it is not surprising that the addition of other alloying elements to copper steels has so little effect on the properties.

Normalized and Drawn Low Alloy Copper Cast Steel.

17. Some of the cast steels listed in Table 4 were double normalized and drawn and their properties are shown in Table 5.

18. Steels containing copper and vanadium, titanium, aluminum, molybdenum or chromium had equal or slightly better strength values and notably better impact resistance than the plain

Table 4

MECHANICAL PROPERTIES OF FULLY ANNEALED CAST STEELS CONTAINING COPPER WITH OTHER ALLOYS

Chemical Composition, Per Cent							Mechanical Properties				
C	*Mn*	*Si*	*P*	*S*	*Cu*	*Other Elements*	*Tensile Strength lb. per sq. in.*	*Yield Strength lb. per sq. in.*	*Elongation in 2 in., Per Cent*	*Reduction of Area, Per Cent*	*Izod Impact ft.-lb.*
0.19	0.63	0.33	.....	.....	.....	.....	62,000	42,000	36.5	59.8	61.0
0.25	0.68	0.36	.....	.....	.....	.....	66,000	40,000	32.5	49.8	38.0
0.32	0.80	0.40	.....	.....	.....	.....	71,500	42,000	30.0	45.5	26.0
0.23	0.63	0.33	0.011	0.035	1.27	.....	79,500	57,500	27.8	50.3	30.0
0.29	0.66	0.34	0.014	0.032	1.24	.....	85,000	57,000	25.8	44.3	22.5
0.17	0.67	1.05	0.012	0.029	1.25	.....	80,000	60,500	30.8	56.5	36.5
0.17	1.19	1.04	0.012	0.032	1.26	.....	83,500	61,500	29.8	57.0	38.5
0.18	0.60	0.31	0.010	0.036	1.28	0.15 V	81,000	58,000	28.5	54.1	34.0
0.26	0.67	0.33	0.009	0.037	1.28	0.15 V	85,500	61,000	27.3	50.2	27.5
0.18	1.18	0.37	0.013	0.032	1.24	.....[1]	79,500	56,500	30.3	56.0	38.0
0.25	1.21	0.37	0.010	0.031	1.19	.....	84,000	57,000	26.6	48.1	14.0
0.32	1.52	0.40	0.014	0.031	1.28	.....	96,000	61,500	24.0	43.4	6.0
0.16	0.62	0.33	0.010	0.029	1.27	0.05 Ti*	76,500	60,500	29.8	56.0	37.5
0.19	0.67	0.36	0.012	0.030	1.26	0.10 Ti*	76,000	56,000	30.5	55.2	37.0
0.25	0.66	0.35	0.013	0.029	1.26	0.05 Ti*	81,500	60,000	29.5	50.6	36.0
0.23	0.69	0.38	0.013	0.029	1.26	0.10 Ti*	78,000	56,500	29.0	52.5	33.5
0.17	0.65	0.35	0.014	0.030	1.25	0.10 Al*	75,000	55,500	30.8	53.2	43.0
0.20	0.63	0.35	0.009	0.032	1.25	0.10 Al*	76,500	55,500	29.8	49.9	37.0
0.18	0.81	0.44	0.011	0.026	1.26	0.20 Mo	77,000	58,000	30.5	55.0	28.5
0.22	0.71	0.33	0.012	0.027	1.28	0.19 Mo	79,500	59,500	28.0	51.7	28.0
0.17	0.59	0.37	0.012	0.028	1.23	0.20 Cb*	78,500	57,000	24.5	41.0	20.0
0.22	0.62	0.37	0.011	0.026	1.20	0.20 Cb*	80,000	61,000	24.0	40.2	18.0
0.15	0.61	0.34	0.010	0.031	1.26	0.10 Zr*	74,500	56,500	29.8	56.3	41.5
0.21	0.61	0.34	0.010	0.028	1.27	0.10 Zr*	77,500	54,500	29.3	50.9	34.0
0.17	0.65	0.34	0.010	0.028	1.25	0.46 Cr	79,000	56,000	28.3	55.9	45.5
0.26	0.64	0.35	0.009	0.029	1.29	0.47 Cr	86,000	55,500	25.0	45.6	30.0

* Per cent added to melt.
[1] Heated at 1600°F. and furnace cooled. All others heated 1 hr. at 1650°F. and then furnace cooled.

Table 5

MECHANICAL PROPERTIES OF NORMALIZED AND NORMALIZED AND DRAWN COPPER CAST STEELS CONTAINING OTHER ALLOYING ELEMENTS

Chemical Composition, Per Cent							Heat Treatment			Mechanical Properties				
C	Mn	Si	P	S	Cu	Other Elements	Normalized °F.	°F.	Drawn °F.	Tensile Strength, lb. per sq. in.	Yield Strength, lb. per sq. in.	Elongation in 2 in., Per Cent	Reduction of Area, Per Cent	Izod Impact ft.-lb.
0.19	0.63	0.33	.....	.....	.....	.....	1700	1600	1200	63,500	44,000	39.0	67.0	64.0
0.28	0.64	0.34	.....	.....	.....	.....	1700	1600	1200	69,000	43,500	37.8	63.3	45.5
0.23	0.63	0.33	.....	.....	1.27	.....	1700	1600	1200	75,500	58,500	33.0	63.9	45.5
0.29	0.66	0.34	.....	.....	1.24	.....	1700	1600	1200	80,000	58,000	30.3	54.7	36.0
0.18	0.61	0.31	.....	.....	1.28	0.15 V	1700	1600	1200	78,500	64,000	31.0	64.6	49.5
0.25	0.65	0.33	.....	.....	1.28	0.15 V	1700	1600	1200	88,500	71,500	26.8	54.9	41.0
0.16	0.62	0.33	.....	.....	1.27	0.05 Ti*	1700	1600	1200	69,000	55,500	34.8	66.3	78.0
0.25	0.66	0.35	.....	.....	1.26	0.05 Ti*	1700	1600	1200	79,000	62,500	30.3	58.2	51.5
0.19	0.67	0.36	.....	.....	1.26	0.10 Ti*	1700	1600	1200	75,000	63,000	29.8	60.4	59.5
0.23	0.69	0.38	.....	.....	1.26	0.10 Ti*	1700	1600	1200	80,500	64,500	29.5	61.8	54.5
0.17	0.65	0.35	.....	.....	1.25	0.10 Al*	1700	1600	1200	72,500	58,500	32.8	59.6	55.0
0.20	0.63	0.35	.....	.....	1.25	0.10 Al*	1700	1600	1200	75,000	60,000	32.5	60.1	53.5
0.18	0.81	0.44	.....	.....	1.26	0.20 Mo	1700	1600	1200	73,500	58,500	34.3	68.1	67.0
0.22	0.71	0.33	.....	.....	1.28	0.19 Mo	1700	1600	1200	74,500	59,000	32.0	61.0	68.5
0.17	0.59	0.37	.....	.....	1.23	0.20 Cb*	1700	1600	1200	78,500	67,000	23.5	42.2	29.0
0.22	0.62	0.37	.....	.....	1.20	0.20 Cb*	1700	1600	1200	82,500	69,500	22.3	38.8	26.5
0.15	0.61	0.34	.....	.....	1.26	0.10 Zr*	1700	1600	1200	70,500	58,500	33.3	67.3	66.5
0.21	0.61	0.34	.....	.....	1.27	0.10 Zr*	1700	1600	1200	75,000	58,500	29.5	55.4	39.0
0.17	0.65	0.34	.....	.....	1.25	0.46 Cr	1700	1600	1200	76,500	59,500	31.3	66.0	68.5
0.26	0.64	0.35	.....	.....	1.29	0.47 Cr	1700	1600	1200	84,500	59,000	28.8	59.4	52.0
0.30	0.82	0.33	.....	.....	1.14	0.77 Cr	1700	1575	.....	110,000	70,250	20.5	47.5	12.5
							1700	1572	750	113,250	73,250	22.7	52.9	8.0
							1700	1575	1050	121,000	84,000	20.0	49.1	8.0

* Amount added to steel. No analysis made for residual alloy content.
All steels were heated for 1 hr. at the temperatures indicated.

copper steels. In addition, vanadium, titanium, and aluminum raised the yield strength.

19. While columbium in the copper cast steels markedly raised the yield strength, it lowered the ductility and impact resistance perceptibly.

20. The properties of the steel containing 0.77 per cent chromium illustrated the necessity for a high draw treatment, for without this treatment the ductility and resistance to impact was low.

21. Table 6 was prepared to demonstrate the influence of small amounts of titanium and aluminum on properties of normalized and normalized and drawn copper steels. The principal effect of adding from 0.05 to 0.10 per cent of these elements to the steels was to raise the yield strength and resistance to impact. Addition of 0.20 per cent titanium in the manganese-copper steels further increased the yield strength, but decreased the Izod impact and ductility slightly.

Properties of Normalized Copper Cast Steels Containing Silicon and Manganese

22. Table 7 summarizes the properties of several normalized copper steels containing more than the usual percentages of manganese and silicon. Increasing the silicon content up to 1 per cent appeared to have no effect on the properties of the low carbon copper steels of normal manganese content, but in steels with manganese contents of from 1.00 to 1.20 per cent both tensile and yield strengths were raised. Neither the excess silicon nor the manganese affected the response of the copper steels to the precipitation hardening treatment.

23. Finlayson[6] has shown that commercial copper-manganese-silicon cast steels in the fully annealed condition have proved satisfactory under very severe service conditions.

Properties of Normalized Copper-Manganese-Titanium Cast Steels

24. From previous work it appeared that the copper-manganese-titanium steels would have properties suitable for low alloy, high-tensile steel castings. Manganese and copper supplement each other in conferring high tensile and yield strengths to cast steels. Titanium was found to reduce the tendency for these higher manganese steels to air harden and to improve the yield strength,

Table 6

EFFECT OF ALUMINUM AND TITANIUM ON PROPERTIES OF COPPER CAST STEELS

Chemical Composition, Per Cent							Heat Treatment			Mechanical Properties				
C	Mn	Si	P	S	Cu	Other Elements	Normalized °F.	°F.	Drawn °F.	Tensile Strength, lb. per sq. in.	Yield Strength, lb. per sq. in.	Elongation in 2 in., Per Cent	Reduction of Area, Per Cent	Izod Impact ft.-lb.
0.32	0.78	0.37	0.009	0.028	1.22	.....	1650	.....	.....	96,500	64,500	25.0	49.7	20.0
0.29	0.66	0.34	0.014	0.032	1.24	.....	1700	1600	1200	80,000	58,000	30.3	54.7	36.0
0.30	0.77	0.41	0.012	0.026	1.20	0.10 Ti*	1700	1600	.....	96,700	68,900	22.0	40.0	36.0
0.31	0.80	0.37	0.012	0.025	1.27	0.20 Ti*	1700	1600	.....	94,200	71,300	23.0	42.0	45.0
							1700	1600	1200	85,300	64,000	28.0	60.0	48.1
0.32	0.83	0.40	0.011	0.024	1.29	0.10 Al*	1700	1600	.....	97,900	71,800	24.0	45.0	44.0
							1700	1600	1200	87,500	64,000	26.0	51.0	41.0
MEDIUM MANGANESE COPPER STEELS														
0.27	1.15	0.30	0.024	0.021	1.10	.....	1600	.....	700	96,250	60,750	27.2	57.8	36.5
0.28	1.11	0.32	.....	.....	1.10	0.05 Ti*	1600	.....	700	96,000	67,500	28.0	55.5	48.0
0.29	1.30	0.31	0.024	0.020	1.10	.....	1600	.....	700	102,250	66,600	25.7	55.7	38.0
0.29	1.29	0.31	.....	.....	1.10	0.05 Ti*	1600	.....	700	102,500	71,250	26.7	53.2	39.2
0.27	1.34	0.38	0.038	0.042	1.20	0.20 Ti*	1600	.....	750	104,000	81,000	25.5	52.8	30.0
0.28	1.09	0.35	0.038	0.042	1.02	0.20 Ti*	1600	.....	750	99,000	75,250	26.0	49.2	34.5

* These values represent the amount added to the steels. Amount recovered not determined.

All steels were heated for 1 hr. at the temperatures indicated.

Table 7

EFFECT OF SILICON AND MANGANESE ON PROPERTIES OF NORMALIZED COPPER CAST STEELS

Chemical Composition, Per Cent						Heat Treatment		Mechanical Properties					
C	Mn	Si	P	S	Cu	Normalized °F.	Drawn °F.	Tensile Strength, lb. per sq. in.	Yield Strength, lb. per sq. in.	Elongation in 2 in., Per Cent	Reduction of Area, Per Cent	Impact* ft.-lb.	Brinell Hardness
LOW CARBON STEELS													
0.19	0.60	0.44	0.028	0.031	.....	1650	.....	71,750	46,500	34.0	57.9	24.0 I	139
0.19	0.63	0.35	.....	.....	1.27	1650	1575	86,250	64,000	28.0	50.0	.....	170
0.17	0.68	0.64	.....	.....	1.32	1750	.....	79,500	67,000	29.5	46.4	22.5 C	144
0.17	0.68	0.64	.....	.....	1.32	1750	840[1]	85,000	72,000	26.7	46.6	22.0 C	162
0.15	0.65	0.76	.....	.....	1.32	1750	.....	81,000	67,000	29.7	51.4	27.0 C	151
0.15	0.65	0.76	.....	.....	1.32	1750	840[1]	94,000	81,000	24.7	44.8	18.0 C	179
0.15	0.65	0.94	.....	.....	1.32	1750	.....	82,000	68,000	31.2	52.3	28.0 C	152
0.15	0.65	0.94	.....	.....	1.32	1750	840[1]	96,000	79,000	27.5	51.9	18.0 C	176
0.17	0.67	1.05	0.012	0.029	1.25	1700	1600 1200	80,000	64,000	32.3	64.3	53.0 I	.....
0.13	0.53	1.02	.....	.....	1.24	1750	.....	79,000	71,000	22.5	29.2	23.0 C	151
0.18	1.18	0.37	.....	.....	1.24	1700	1600 1200	77,500	58,500	30.5	62.8	56.0 I	.....
0.17	1.19	1.04	.....	.....	1.26	1700	1600 1200	85,500	66,000	29.5	59.6	43.5 I	.....
0.17	1.03	1.07	.....	.....	1.28	1750	.....	89,000	65,000	29.0	55.7	20.2 C	164
0.17	1.03	1.07	.....	.....	1.28	1750	840[1]	102,000	77,000	22.5	40.7	9.0 C	197
0.18	1.00	1.70	.....	.....	1.28	1750	.....	98,000	71,000	27.5	50.0	15.0 C	185
0.18	1.00	1.70	.....	.....	1.28	1750	840[1]	107,000	92,000	24.5	46.3	11.5 C	208
MEDIUM CARBON STEELS													
0.22	0.56	0.61	.....	.....	1.16	1750	.....	82,000	61,000	30.5	55.0	23.5 C	150
0.22	0.56	0.61	.....	.....	1.16	1750	840[1]	90,500	70,500	27.7	49.5	20.0 C	166
0.27	1.15	0.30	0.024	0.021	1.10	1600	.....	95,750	60,750	26.0	54.0	33.5 I	.....
0.29	1.30	0.31	0.024	0.020	1.10	1600	.....	102,500	67,000	24.5	51.9	32.5 I	.....
0.30	0.89	1.14	.....	.....	1.16	1750	.....	99,000	72,000	26.0	48.0	17.5 C	184
0.30	0.89	1.14	.....	.....	1.16	1750	840[1]	109,000	84,000	23.0	38.8	9.0 C	197
0.34	1.17	0.32	0.024	0.021	1.10	1600	.....	104,000	69,000	24.0	44.6	30.0 I	.....
0.34	1.33	0.31	0.024	0.020	1.12	1600	.....	110,250	70,500	23.2	50.1	24.0 I	.....

[1] Reheated for 3 hours at 840°F.; then air cooled. All others were heated for 1 hr. at the temperatures indicated.

* C = Charpy, I = Izod.

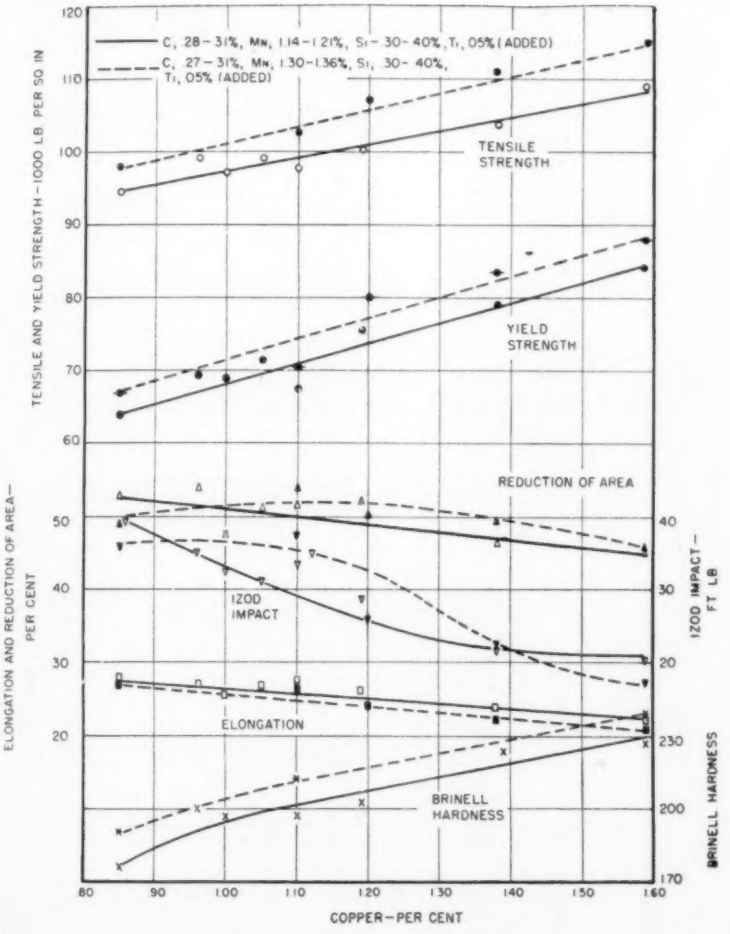

C, 28–31%, Mn, 1.14–1.21%, Si– 30–40%, Ti, 05% (ADDED)
C, 27–31%, Mn, 1.30–1.36%, Si, 30–40%, Ti, 05% (ADDED)
TENSILE STRENGTH
YIELD STRENGTH
REDUCTION OF AREA
IZOD IMPACT
ELONGATION
BRINELL HARDNESS
TENSILE AND YIELD STRENGTH – 1000 LB PER SQ IN
ELONGATION AND REDUCTION OF AREA – PER CENT
IZOD IMPACT – FT LB
BRINELL HARDNESS
COPPER – PER CENT

FIG. 5—MECHANICAL PROPERTIES OF NORMALIZED COPPER-MANGANESE-TITANIUM CAST STEELS

ductility and impact resistance. The mechanical properties of some of the typical compositions studied are summarized in Table 8.

25. The changes in properties with changes in copper and manganese contents are shown in Fig. 5. Curves in this figure show that the tensile and yield strengths and the Brinell hardness increase proportionally while the ductility and impact resistance decrease with copper content between 0.80 and 1.60 per cent. The distance between the solid and dotted curves for the individual

Table 8

EFFECTS OF COPPER, CARBON, MANGANESE, AND TITANIUM ON MECHANICAL PROPERTIES OF CAST STEELS

Chemical Composition, Per Cent							Heat Treatment			Mechanical Properties					
C	Mn	Si	P	S	Cu	Ti*	Normalized °F.	°F.	Drawn °F.	Tensile Strength, lb. per sq. in.	Yield Strength, lb. per sq. in.	Elongation, in 2 in., Per Cent	Reduction of Area, Per Cent	Izod Impact ft.-lb.	Brinell Hardness
LOW MANGANESE STEELS															
0.23	0.69	0.38	0.013	0.029	1.26	0.10	1700	1600	1200	80,500	64,500	29.5	61.8	54.5	.....
0.31	0.80	0.37	0.012	0.025	1.27	0.20	1700	1600	.....	94,250	71,500	23.0	42.0	45.0	.....
							1700	1600	1200	85,250	64,000	28.0	60.0	48.0	.....
MEDIUM MANGANESE STEELS															
0.22	1.24	0.37	0.037	0.040	0.89	0.05	1600	.....	.....	88,250	63,250	28.7	56.0	45.0	.....
							1600	.....	1150	87,250	64,750	29.7	57.4	35.5	.....
0.22	1.35	0.37	0.040	0.041	0.87	0.05	1600	.....	.....	92,500	66,000	29.7	57.0	37.5	.....
							1600	.....	1150	89,500	65,500	28.7	56.8	34.5	.....
0.23	1.14	0.39	0.037	0.040	0.85	0.05	1600	.....	.....	89,500	61,000	29.5	55.4	42.5	.....
							1600	.....	1150	90,000	68,750	28.5	56.0	33.5	.....
0.23	1.16	0.34	0.037	0.040	1.00	0.05	1600	.....	.....	91,500	65,250	30.0	57.1	46.0	.....
							1600	.....	1150	91,000	72,000	28.0	58.3	38.0	.....
0.23	1.30	0.37	0.046	0.042	1.04	0.05	1600	.....	.....	93,500	66,500	28.0	55.3	41.0	.....
							1600	.....	1150	90,750	68,500	26.0	57.0	35.0	.....
0.23	1.21	0.33	0.037	0.039	1.14	0.05	1600	.....	.....	91,500	64,500	29.0	54.8	36.5	.....
							1600	.....	1150	90,750	70,250	27.0	57.6	32.5	.....
0.23	1.13	0.29	0.039	0.041	0.99	0.20	1600	.....	.....	91,250	68,250	28.5	57.5	45.0	.....
							1600	.....	750	91,250	68,500	28.5	56.5	44.0	.....
							1600	.....	1050	100,250	78,750	25.7	52.9	24.0	.....
0.27	1.31	0.37	0.036	0.039	0.85	0.05	1600	.....	.....	98,000	67,000	27.0	49.1	36.0	187-192
							1600	.....	1150	96,000	68,500	27.0	55.3	30.5	187-197
0.27	1.34	0.38	0.038	0.042	1.20	0.20	1600	.....	.....	104,000	79,000	25.5	53.8	29.0	.....
							1600	.....	750	104,000	81,000	25.5	52.8	30.0	.....
							1600	.....	1050	106,750	86,000	23.0	50.5	20.0	.....
0.28	1.17	0.38	0.036	0.040	0.85	0.05	1600	.....	.....	94,500	64,000	28.2	53.0	39.5	177
							1600	.....	1150	93,500	67,500	28.0	57.2	35.5	187
0.28	1.11	0.32	0.024	0.021	1.10	0.05	1600	.....	700	96,000	67,500	28.0	55.5	48.0	192-197

(Continued on next page)

Table 8 *(Continued)*

0.28	1.22	0.38	0.040	0.041	1.10	0.05	1500	.....	.....	98,250	68,750	25.2	46.9	34.0	.....
							1600	.....	.....	97,750	67,750	26.7	51.4	30.0	.....
							1700	.....	.....	98,000	68,500	25.5	51.2	29.0	.....
							1800	.....	.....	105,000	62,500	24.5	51.2	7.5	.....
0.28	1.18	0.40	0.040	0.040	1.05	0.05	1600	.....	.....	99,000	71,500	26.7	51.4	30.5	.....
							1600	.....	1100	102,000	77,750	24.7	50.1	20.0	201-207
							1600	.....	1150	97,250	72,500	25.0	51.0	25.5	194-199
							1600	.....	1250	88,750	65,750	28.7	56.3	32.0	179-185
0.31	1.12	0.38	0.041	0.041	1.00	0.05	1600	.....	.....	97,000	69,000	25.2	47.4	32.5	190-197
							1600	.....	1050	104,500	79,750	24.0	50.8	17.0	217-218
							1700	1575	.....	97,000	71,250	27.0	52.0	37.5	.....
							1700	1575	1050	104,250	81,750	23.2	48.1	17.0	.....
0.29	1.17	0.37	0.042	0.042	1.19	0.05	1600	.....	.....	100,000	75,500	26.0	52.2	28.5	197-204
							1600	.....	1050	107,000	85,500	23.3	49.8	17.0	223-229
0.31	1.36	0.31	0.041	0.042	1.20	0.05	1600	.....	.....	107,500	80,000	23.7	50.4	26.0	.....
							1600	.....	1050	109,250	85,500	23.0	50.0	17.0	.....
							1700	1575	.....	107,500	82,000	24.7	49.6	31.0	.....
							1700	1575	1050	109,500	83,750	23.5	51.7	18.0	.....
0.29	1.18	0.30	0.042	0.043	1.38	0.05	1600	.....	.....	103,750	79,000	23.7	46.2	22.5	212-223
							1600	.....	1050	106,000	83,500	23.3	50.0	17.5	217-223
0.33	1.34	0.32	0.041	0.050	1.38	0.05	1600	.....	.....	111,000	83,750	22.5	49.5	22.0	.....
							1600	.....	1050	110,000	86,000	23.2	50.8	17.5	.....
0.30	1.17	0.33	0.041	0.042	1.59	0.05	1600	.....	.....	109,250	84,250	22.5	45.0	20.0	226
							1600	.....	1050	109,000	86,500	22.5	49.2	15.0	.....
0.30	1.34	0.32	0.042	0.042	1.59	0.05	1600	.....	.....	115,000	88,000	21.2	45.5	17.0	235-239
							1600	.....	1050	110,000	87,500	22.7	50.3	14.0	229-232
							1700	1575	.....	112,500	87,500	21.5	46.7	19.5	.....
							1700	1575	1050	109,250	87,850	23.2	50.8	19.0	.....

* Per cent Ti added as medium carbon ferro-carbon-titanium. Amount recovered not determined.
Steels were heated for 1 hr. at the indicated temperatures.

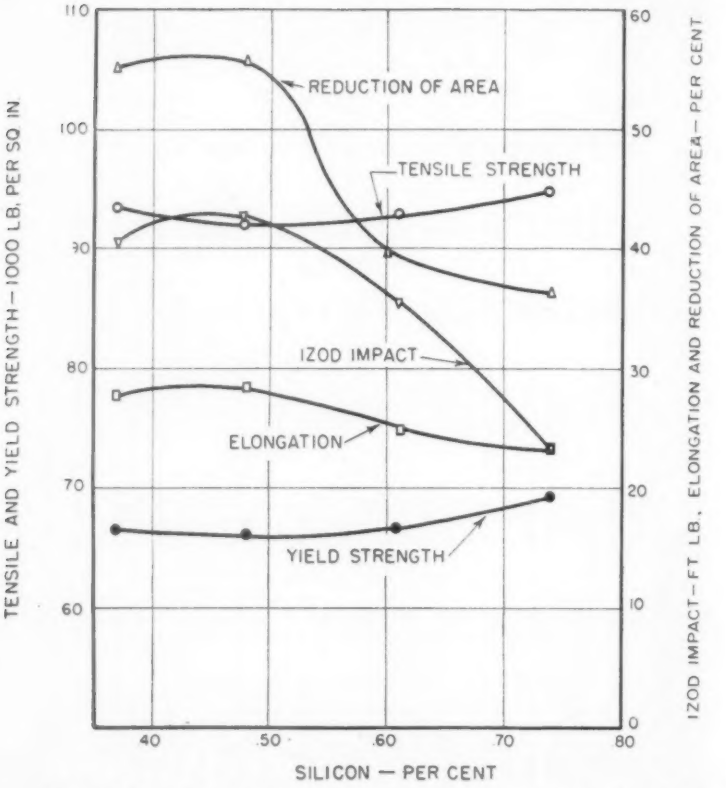


FIG. 6—EFFECT OF SILICON ON STEELS CONTAINING 0.23—0.25 PER CENT C, 1.14—1.30 PER CENT MN, 0.035—0.041 PER CENT P AND S, 1.04—1.14 PER CENT CU, 0.05 PER CENT TI (AMOUNT ADDED) NORMALIZED AT 1600°F.

properties represents the change in properties as a result of a change in manganese content from approximately 1.15 to 1.35 per cent.

26. Table 9 shows the effect of rather high silicon contents on the copper-manganese-titanium cast steels. The relationship of mechanical properties of the steels normalized at 1600°F. to the silicon content is shown in Fig. 6. The tensile and yield strengths are unaffected but the ductility and impact resistance drop off rapidly when the silicon content exceeds 0.50 per cent.

27. One of the effects of titanium in the copper-manganese-titanium steel is to decrease the grain size and to cause grain refinement after heat treating.

Table 9

EFFECT OF SILICON ON PROPERTIES OF COPPER-MANGANESE-TITANIUM STEELS

Chemical Composition, Per Cent							Heat Treatment			Mechanical Properties				
							Normalized		Drawn	Tensile Strength,	Yield Strength,	Elongation, in 2 in.,	Reduction of Area,	Izod Impact
C	Mn	Si	P	S	Cu	Ti*	°F.	°F.	°F.	lb. per sq. in.	lb. per sq. in.	Per Cent	Per Cent	ft.-lb.
0.23	1.21	0.33	0.037	0.039	1.14	0.05	1600	.....	.....	91,500	64,500	29.0	54.8	36.5
							1600	.....	1150	90,750	70,250	27.0	57.6	32.5
0.23	1.16	0.48	0.036	0.039	1.04	0.05	1600	.....	.....	92,000	66,000	28.7	55.9	42.5
							1600	.....	1150	92,000	71,500	26.5	55.6	32.5
0.23	1.16	0.61	0.036	0.043	1.06	0.05	1600	.....	.....	93,000	66,500	25.0	39.8	36.0
							1600	.....	1150	94,000	72,500	23.7	42.0	27.0
0.24	1.23	0.69	0.046	0.041	1.11	0.05	1600	.....	.....	101,500	76,000	21.2	32.9	22.5
							1600	.....	1150	98,000	76,000	19.5	30.5	16.5
0.25	1.14	0.74	0.036	0.038	1.04	0.05	1600	.....	.....	95,000	69,000	23.5	36.3	29.5
							1600	.....	1150	95,500	75,000	21.0	35.2	23.0
0.27	1.20	0.80	0.036	0.043	0.97	0.10	1600	.....	.....	100,000	77,500	20.7	24.7	19.0
							1600	.....	1050	107,000	87,000	17.0	24.4	12.0

* Per cent titanium added. Amount recovered not determined.
Steels were heated for 1 hr. at the indicated temperatures.

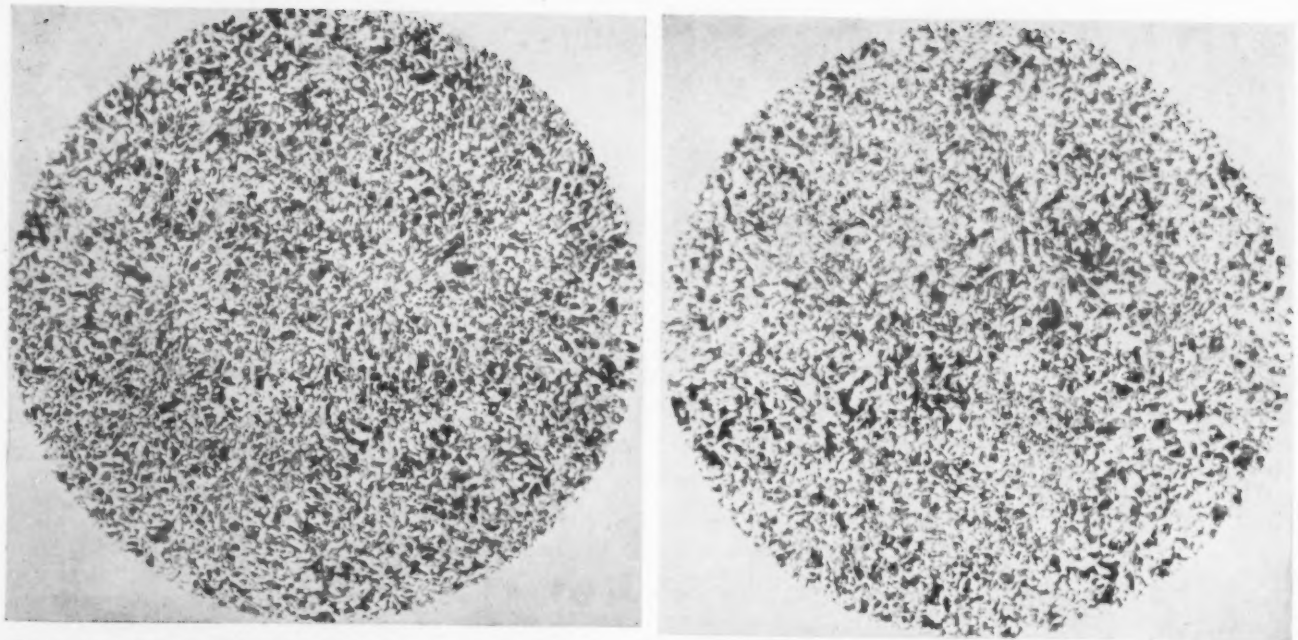
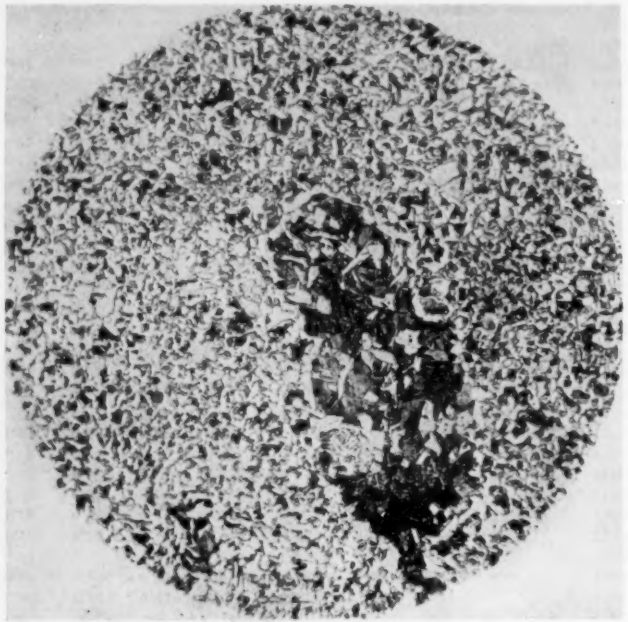


FIG. 7—LEFT—MICROGRAPH OF STEEL CONTAINING 0.28 PER CENT C, 1.22 PER CENT MN, 0.38 PER CENT SI, 0.040 PER CENT P, 0.040 PER CENT S, 1.10 PER CENT CU, 0.05 PER CENT TI (ADDED). ETCHED HNO_3 X 100. NORMALIZED AT 1500°F. TENSILE STRENGTH 98,250 LB. PER SQ. IN. YIELD STRENGTH 68,750 LB. PER SQ. IN. ELONGATION IN 2 IN. 25.2 PER CENT. REDUCTION OF AREA 46.9 PER CENT. IZOD IMPACT 34.0 FT.-LB.

FIG. 8—RIGHT—MICROGRAPH OF STEEL CONTAINING 0.28 PER CENT C, 1.22 PER CENT MN, 0.38 PER CENT SI, 0.040 PER CENT P, 0.040 PER CENT S, 1.10 PER CENT CU, 0.05 PER CENT TI (ADDED). ETCHED HNO_3 X 100. NORMALIZED AT 1600°F. TENSILE STRENGTH 97,750 LB. PER SQ. IN. YIELD STRENGTH 67,750 LB. PER SQ. IN. ELONGATION IN 2 IN. 26.7 PER CENT. REDUCTION OF AREA 51.4 PER CENT. IZOD IMPACT 30.0 FT.-LB.



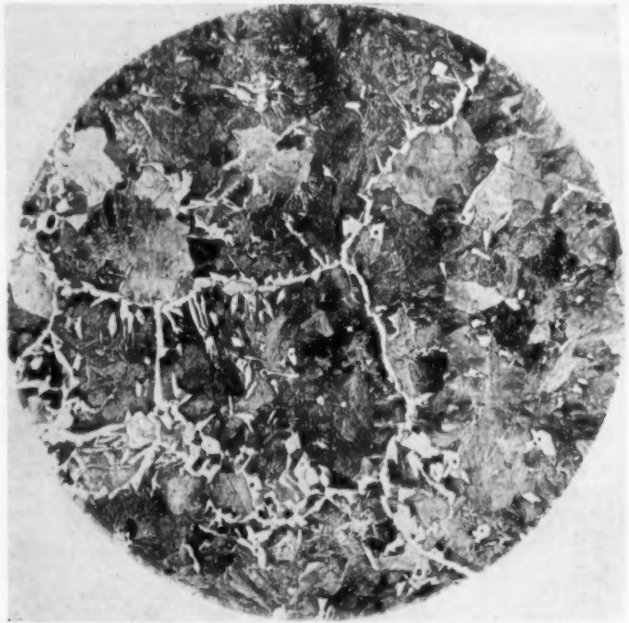


FIG. 9—LEFT—MICROGRAPH OF STEEL CONTAINING 0.28 PER CENT C, 1.22 PER CENT MN, 0.38 PER CENT SI, 0.040 PER CENT P, 0.040 PER CENT S, 1.10 PER CENT CU, 0.05 PER CENT TI (ADDED). ETCHED HNO_3 X 100. NORMALIZED AT 1700°F. TENSILE STRENGTH 98,000 LB. PER SQ. IN. YIELD STRENGTH 68,500 LB. PER SQ. IN. ELONGATION IN 2 IN. 25.5 PER CENT. REDUCTION OF AREA 51.2 PER CENT. IZOD IMPACT 29.0 FT.-LB.

FIG. 10—RIGHT—MICROGRAPH OF STEEL CONTAINING 0.28 PER CENT C, 1.22 PER CENT MN, 0.38 PER CENT SI, 0.040 PER CENT P, 0.040 PER CENT S, 1.10 PER CENT CU, 0.05 PER CENT TI (ADDED). ETCHED HNO_3 X 100. NORMALIZED AT 1800°F. TENSILE STRENGTH 105,000 LB. PER SQ. IN. YIELD STRENGTH 62,500 LB. PER SQ. IN. ELONGATION IN 2 IN. 24.5 PER CENT. REDUCTION OF AREA 51.2 PER CENT. IZOD IMPACT 7.5 FT.-LB.

Table 10

MECHANICAL PROPERTIES OF NORMALIZED AND NORMALIZED AND DRAWN COPPER-CHROMIUM CAST STEELS

Chemical Composition, Per Cent								Heat Treatment			Mechanical Properties				
C	Mn	Si	P	S	Cu	Cr	Other* Elements	Normalized °F.	°F.	Drawn °F.	Tensile Strength, lb. per sq. in.	Yield Strength, lb. per sq. in.	Elongation in 2 in., Per Cent	Reduction of Area, Per Cent	Izod Impact ft.-lb.
0.26	0.64	0.35	0.009	0.029	1.29	0.47	.....	1700	1600	1200	84,500	59,000	28.8	59.4	52.0
0.29	0.75	0.33	0.040	0.041	1.20	0.52	0.05 Ti	1700	1575	.....	99,000	72,000	23.2	41.4	26.0
								1700	1575	1050	110,250	84,250	22.2	48.6	11.0
0.26	0.74	0.39	0.013	0.027	1.25	0.50	0.10 Ti	1700	1600	.....	96,500	69,800	26.0	57.0	45.0
0.26	0.79	0.38	0.014	0.026	1.28	0.52	0.10 Ti	1700	1600	1200	88,200	63,600	28.0	62.0	57.0
0.26	0.78	0.36	0.012	0.024	1.27	0.54	0.20 Ti	1700	1600	.....	95,200	71,100	26.0	58.0	58.0
0.24	0.79	0.38	0.011	0.023	1.28	0.55	0.10 Al	1700	1600	.....	94,800	70,200	26.0	54.0	54.0
0.24	0.79	0.38	0.011	0.023	1.28	0.55	0.10 Al	1700	1600	1200	87,400	63,000	27.0	57.0	58.0
0.32	0.80	0.47	0.012	0.024	1.20	0.53	0.10 Ti	1700	1600	.....	105,200	72,900	24.0	54.0	39.0
0.33	0.81	0.35	0.011	0.024	1.26	0.54	0.10 Al	1700	1600	.....	104,600	71,300	22.0	48.0	39.0
0.33	0.81	0.35	0.011	0.024	1.26	0.54	0.10 Al	1700	1600	1200	93,900	64,000	24.0	51.0	44.0
0.32	0.82	0.38	0.012	0.023	1.27	0.55	0.20 Al	1700	1600	.....	104,300	74,400	22.0	47.0	43.0
0.29	0.79	0.31	0.037	0.041	1.14	0.76	0.05 Ti	1700	1575	.....	105,250	72,000	24.0	54.7	9.7
								1700	1575	1050	113,750	83,000	22.0	50.6	8.5
0.29	0.76	0.33	0.037	0.040	1.15	0.77	0.10 Ti	1700	1575	.....	103,750	76,000	25.0	52.2	17.0
								1700	1575	1050	118,000	85,250	21.7	49.3	13.0
0.24	0.79	0.30	0.035	0.040	1.00	0.69	0.20 Ti	1700	1575	.....	93,750	69,000	26.0	52.5	41.0
								1700	1575	1050	104,750	82,000	23.7	51.1	19.5
0.31	0.79	0.34	0.037	0.041	1.17	0.78	0.05 Al	1700	1575	.....	106,000	76,250	21.5	40.7	25.5
								1700	1575	1050	118,250	91,000	18.2	34.6	10.0
0.31	0.79	0.32	0.036	0.040	1.14	0.78	0.10 Al	1700	1575	.....	105,000	75,750	19.5	33.8	30.0
								1700	1575	1050	116,750	92,600	18.2	35.7	14.0
0.26	0.79	0.33	0.037	0.041	1.00	0.67	0.20 Al	1700	1575	.....	94,000	68,750	24.5	42.5	35.0
								1700	1575	1050	106,250	84,500	23.0	47.8	15.5
0.26	0.79	0.37	0.012	0.025	1.27	1.02	.....	1700	1600	.....	122,250	82,750	18.0	48.0	13.0
								1700	1600	1200	102,100	62,800	24.0	62.0	38.0
0.29	0.81	0.35	0.011	0.024	1.26	1.04	0.10 Ti	1700	1600	.....	115,250	73,000	22.0	56.0	31.0
								1700	1600	1200	107,000	76,750	22.0	58.0	64.0

* Percentage of aluminum or titanium added. Titanium was added as medium carbon ferro-carbon-titanium. Amount recovered not determined. Steels were heated for 1 hr. at the temperatures indicated.

28. To show the effects of various normalizing temperatures on grain size and properties, a steel containing 0.28 per cent C, 1.22 per cent Mn, 0.38 per cent Si, 0.04 per cent P, 0.41 per cent S, 1.10 per cent Cu and 0.05 per cent Ti (amount of Ti added) was normalized at 1500, 1600, 1700, and 1800°F. for 1 hr. Normalizing at 1500, 1600, and 1700°F. produced very little change in properties; whereas at 1800°F. the tensile strength was increased and the Izod impact value was decreased to a very low value. The microstructures of the steel after the normalizing treatments are shown in Figs. 7 to 10 inclusive. Grain coarsening of this steel takes place on normalizing at a temperature somewhere between 1600 and 1700°F. At 1800°F. it has completely coarsened.

Properties of Normalized and Drawn Copper Chromium Cast Steels

29. To show the result of adding 0.50 to 1.00 per cent of chromium to normalized copper cast steels of normal manganese content the steels listed in Table 10 were tested. In practically all these steels aluminum or titanium in amounts of 0.05 to 0.20 per cent was added. Steels with 0.50-0.55 per cent chromium and with 0.10 per cent aluminum or titanium had properties comparing favorably with those of the copper-manganese-titanium steels in the normalized state. These steels were only mildly air hardening. On the other hand, steels containing 0.75 to 1.0 per cent chromium were more air hardening and required a larger amount of aluminum or titanium, or a draw treatment to give them satisfactory ductility and impact resistance.

Bibliography

1—Kinnear, H. B., *"One Per Cent Copper Steel Has Desirable Physical Qualities,"* Iron Age, vol. 128, 1931, pp. 696-699, 820-824. Also: U. S. Patent 1,607,086, November 16, 1926.

2—Gregg and Daniloff, *"Alloys of Iron and Copper,"* McGraw-Hill Book Company, Inc., New York, 1934.

3—*"Copper in Cast Steel and Iron,"* Copper Development Association, London, England, 1937.

4—Bullens-Battelle, *"Steel and Its Heat Treatment,"* Vol. I and II. Fourth Edition, John Wiley and Sons, Inc., New York, 1938, 1939.

5—Alexander, M., *"Copper-Steel Castings,"* Third Report of the Steel Castings Research Committee, Iron and Steel Institute, Special Report No. 23, 1938, Section III.

6—Finlayson, A., *"Low Alloy Steel Castings in Logging and Road Construction,"* Metals and Alloys, vol. 8, 1937 pp. 239-244.

DISCUSSION

Presiding: J. W. BOLTON, Lunkenheimer Co., Cincinnati, Ohio

MEMBER: Have the authors done any work on the difficulties of welding of plain-copper steels?

MR. GREENIDGE: As we have not investigated the welding of the plain-copper cast steels, I am unable to give specific data on their weldability. Mr. Kinnear's experience with steels containing one per cent copper was that these steels welded very satisfactorily.

MEMBER: Do the authors know of any definite limit to copper content beyond which the welding would not be satisfactory?

MR. GREENIDGE: We have not determined the percentage of copper that makes a steel unsatisfactory for welding. Most commercial steels probably contain less than two per cent of copper, which, I am sure, does not make them unsuited for welding.

MEMBER: I might say that I have tried some copper steels with up to two and a half to three per cent of copper and have tried welding them. I had no difficulty with the two and a half per cent copper steel. I haven't gone as far as testing the welds for mechanical properties. I would like to know if this investigation has brought out any characteristic of the machinability of copper cast steels?

MR. GREENIDGE: No, we really haven't had an opportunity to compare the machinability of copper cast steels with that of other cast steels. I have not heard, however, that anyone is having difficulty in machining copper cast steels.

MEMBER: What was the radius at the bottom of the notch in your Izod specimens?

MR. GREENIDGE: We used the notch recommended by American Society for Testing Materials which has a radius of 0.010 in. at the bottom.

A. F. A. Transaction

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IN accordance with the publication policy established in 1938, the material contained in the foregoing pages, consisting of papers and discussions presented at the 1939 Cincinnati Convention of A.F.A., is the first section of the quarterly *Transactions*.

Succeeding quarterly issues of *Transactions* will contain papers, reports and proceedings of the Cincinnati Convention together with verbatim discussions presented at convention sessions.

The material contained in the quarterly will be compiled into the annual bound volume of A.F.A. TRANSACTIONS, numbered consecutively with former years. The 1939 bound volume will be known as Vol. XLVII.

The bound volume will be sent to A.F.A. Company and Sustaining Members without charge and will be sent to other members only on definite request and on payment of a fee of $3.00 to cover printing and binding costs.

Code of Recommended Good Practices For Metal Cleaning Sanitation

THE new American Foundrymen's Association Code of Recommended Good Practices for Metal Cleaning Sanitation just off the press is the fourth in a series of Recommended Good Practices developed by the A.F.A. Safety and Hygiene Committee.

This code prescribes the necessary fundamental engineering and construction information for the design, installation, operation and maintenance of all metal cleaning equipment and processes. The use of personal protective devices required in metal cleaning operations for the protection of health and safety of persons and the promotion of good housekeeping is also detailed.

The material in this code as in other A.F.A. Safety and Hygiene codes is presented in such a manner that foundries not having engineering staffs as well as those having engineering staffs can quickly and readily understand it.

Price to non-members $2.50 and to members $1.25.

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